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This review report addresses the microstructural and environmental "short" crack studies that have been conducted since 1965. This includes discussions on present understanding of the mechanisms of "short" crack behavior of materials, and microstructurally "short" crack studies including the effects of grain size, grain and phase boundaries, variation of precipitate size, processing technique and stress ratio. In addition, this report discusses environmental influences in the "short" crack regime. Also, a title list of "short" crack related works published so far is included.

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UNIVERSITY OF UTAH/AFOSR

AGING AIRCRAFT PROGRAM

MODELING CORROSION EFFECTS ON STRUCTURAL INTEGRITY

D. W. HOEPPNER, T. GOSWAMI AND C. VENKATESAN

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OUTLINE

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- EXTENSIVE PITTING, CREVICE, FRETTING, EXFOLIATION, FILIFORM AND GALVANIC CORROSION
- INTERGRANULAR CORROSION
- WHITE POWDER DEPOSITION (f RH)
- OTHER DAMAGES

UNIVERSITY OF UTAH/AFOSR

AGING AIRCRAFT PROGRAM

STARTED MAY 1, 1995

OUR RESEARCH AIMS ARE

TO UNDERSTAND PITTING CORROSION AND FRETTING CORROSION EFFECTS ON AIRCRAFT MATERIAL(S)

TO GENERATE RATE DATA

TO MODEL THE PITTING CORROSION FATIGUE BEHAVIOR USING THE PREVIOUS MODEL OF HOEPPNER, NACE 2, (1971), ASTM (1979).

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VERIFICATION OF THIS MODEL WITH DATA

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3. DESIGN OF EXPERIMENTS
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5. TRANSITION TO LONG CRACK
6. LONG CRACK GROWTH
7. CORROSION FATIGUE CRACK GROWTH

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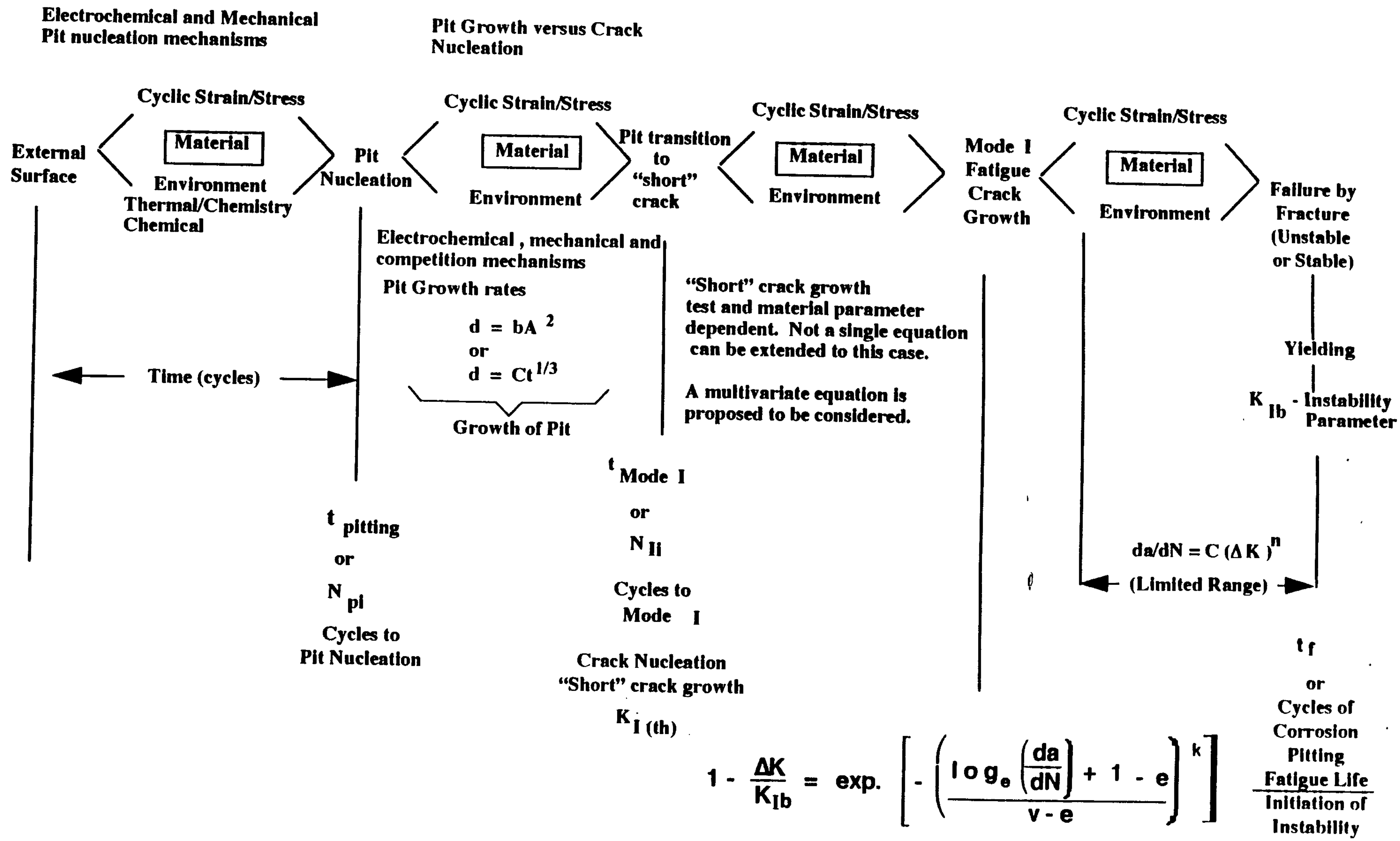
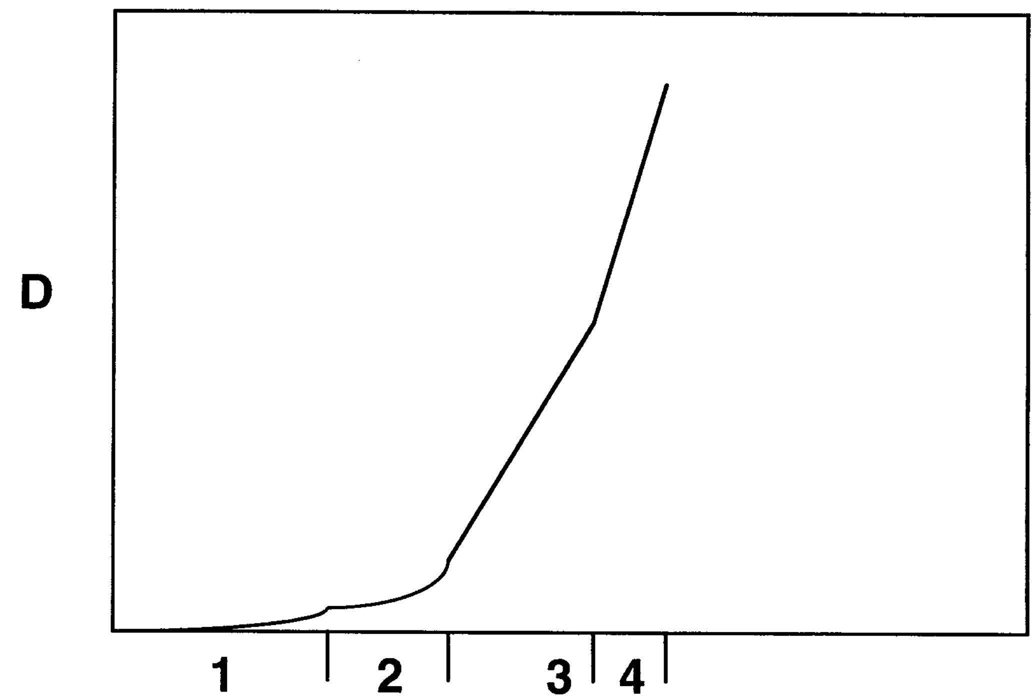


$$1 - \frac{\Delta K}{K_{Ib}} = \exp. \left[- \left(\frac{\log_e \left(\frac{da}{dN}\right) + 1 - e}{v - e} \right)^k \right]$$

DAMAGE GROWTH



1: NUCLEATION AND GROWTH STAGE
2: "SHORT" CRACK GROWTH
3: "LONG" CRACK GROWTH
4: UNSTABLE REGIME

ELECTROCHEMICAL STAGE

ENVIRONMENT
MATERIAL
LOADS
EXTERNAL VARIABLES
INTERACTIONS

PRODUCE PITTING

QUANTIFICATION VERY DIFFICULT - DEVELOPMENT OF METRICS

STARTING POINT - ASTM G 46

PIT GROWTH

ELECTROCHEMICAL
MECHANICAL
MATERIAL
INTERACTIONS

RATE: LINEAR, PARABOLIC, AND CUBIC ETC.

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COMPETITION MECHANISMS

EFFECTS IN PIT GROWTH RATES

CRACK NUCLEATION

EARLY CRACK GROWTH

TRANSITION OF PITS TO CRACKS

"SHORT" CRACKING

MICROSTRUCTURE
CHEMICAL ENVIRONMENT
EMBRITTLEMENT
DEBRIS AND RELAXATION

RATES UNKNOWN

TRANSITION*

FROM A PIT TO A "SHORT" CRACK

"SHORT" CRACK TO "LONG" CRACK

* MORE WILL BE DISCUSSED AT ASME ANNUAL WINTER MEETING AT SAN FRANCISCO ON THE STRUCTURAL INTEGRITY OF AGING AIRCRAFT.

LONG CRACK GROWTH

VERIFY PITTING CORROSION FATIGUE MODEL

UNIVERSITY OF UTAH/AFOSR

AGING AIRCRAFT PROGRAM

SUMMARY

CONTINUE RESEARCH REPORTED

GENERATE MORE DATA IN "SHORT" AND "LONG" CRACK REGIME

VERIFY THE FRAMEWORKS PROPOSED

DEVELOP/MODIFY MODEL(S)

CONTRIBUTE TO THE PRESENT UNDERSTANDING OF TRANSITION FROM A PIT TO A "SHORT" CRACK AND LEFM CRACK

PITTING CORROSION FATIGUE OF STRUCTURAL MATERIALS

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ABSTRACT

Further evaluation of studies related to pitting corrosion fatigue of aircraft structural materials are presented in this paper. This study was undertaken to consider electrochemical effects in pit formation and role of pitting in fatigue and corrosion fatigue crack nucleation behavior. Thus, a review of mechanisms that cause pit nucleation and growth is presented herein. Since the transition of pit(s) to crack(s) is an extremely important issue in assessing the significance of pits on structural integrity the transition models are reviewed and a new conceptual model is presented.

INTRODUCTION:

Pitting corrosion has been found in the structure of a tear-down C/KC 135 aircraft. Widespread pitting on the surface and hidden within the fuselage joints, as found in the case of C/KC 135 and other aircraft components, may pose a significant threat to the structural integrity of an aircraft or a component as pitted regions are those from which fatigue and/or corrosion fatigue cracks may nucleate and propagate. In addition, pitted regions may coalesce thereby forming a longer crack. Aircraft materials, particularly those of high strength types, are susceptible to pitting as well as stress corrosion cracking in a favorable environment (Burleigh, 1991). Since aircraft operate in a spectra of environments, the effect of such environmental influences on fatigue or corrosion fatigue crack growth behavior must be established and ultimately understood.

Early work reported by Hoeppner and Hyler (1966) showed that fatigue life in aluminum alloys improved as the exposure time in a vacuum increased. In a similar study by Bradsaw and Wheeler (1966), crack growth rates were found to vary in the presence of water in the environment or water vapor pressure. In the literature little data have been reported on the role of pitting in fatigue, mechanisms of pit nucleation and growth with respect to crack nucleation and growth, and pitting corrosion fatigue crack growth modeling of aluminum alloys. Hoeppner (1971, 1979) conducted one of the original researches related to pitting and pitting corrosion fatigue modeling in aluminum and other alloys, this has raised the interests of the scientific community and more studies are now underway and results are appearing in the literature. Recent focus on this issue from the KC 135 program has raised the level of interest in these phenomenon as well.

In order to use the potential life of commercial as well as military fleets, knowledge of localized corrosion processes such as pitting and their potential role in integrity of aircraft structures must be gained so that these models can be integrated with the other models to allow accurate assessment of inspection intervals and assure that the high standard of safety will not be affected negatively.

Previous reviews on pit nucleation and growth mechanisms were conducted by Hoeppner and Goswami (1993, 1995a) are expanded in this paper. Corrosion fatigue crack growth behaviors in aluminum and titanium alloys also were compiled by Goswami and Hoeppner (1994, 1995b). The objectives of this research were;

1. To review the mechanisms of pit nucleation and growth.
2. To examine the role of pit growth in the generation of "short corrosion fatigue cracks".
3. To evaluate the "short corrosion fatigue crack growth" behavior until a long crack size is reached.
4. To explore the transition of "short crack" to "long crack" growth where a linear elastic fracture mechanics parameter

such as mode I change in stress intensity (ΔK_I) be applied to model the behavior.

In this paper, particular emphasis is given to the study of fatigue crack nucleation and growth mechanisms from the regions of pits.

REVIEW OF PIT NUCLEATION AND GROWTH MECHANISMS:

The American Society for Metals (1986) defines corrosion "as deterioration of a material due to chemical or electrochemical reaction with its environment". In the pitting corrosion process, local dissolution causes cavities in passivated metals when exposed to solutions containing aggressive anions of chloride types. This is a discontinuous corrosion process. There is a threshold value of anodic potential for a given electrolyte system below which pits do not form. Pit formation and their shapes are random phenomena, in that location and shape of a pit depends upon several material as well as electrochemical factors which are not very well understood. Several pit shapes were shown by Ma and Hoeppner (1994). It has been shown that pits may form on the sites where there is a concentration of constituent particles (or second phases) (Bond et al, 1966 and Zahavi and Yahalom, 1975) since such a site is vulnerable to corrosive attack where the protective film thickness is uneven and even broken, thus allowing local attack. However, how a local dissolution process may penetrate the depth, intergranularly and across the grains has not been established. Clearly, much more research needs to be undertaken to investigate the mechanisms of pitting corrosion and how pit shapes can be controlled so that enhancement of properties of exposed materials of aircraft structures can be accomplished.

Pit Nucleation Mechanisms:

Table 1 provides a summary of pit nucleation mechanisms. They are adsorption related, ion migration and film breakdown mechanisms.

Pit Growth Mechanisms:

The mechanisms of a pit growth is an important phase of activity. During this phase the pit concentrates stresses and localized dissolution removes material. In the pitting corrosion fatigue process, it has been hypothesized that at the early stages only electrochemical parameters influence the generation of pits and their growth. Furthermore, observation of intergranular, randomly oriented cracks frequently have been made inside pits. However, in a latter stage, competition between the pit growth process and the crack nucleation process becomes significant. This, "competition" undoubtedly is determined by energetic processes that need further clarification in order that the development of a mode I (or other appropriate mode crack) can be understood. Therefore, pit growth has a significant contribution to the fatigue crack nucleation and growth mechanisms. This aspect has not yet quantitatively established. Pit growth mechanisms are of the types shown in Table 2.

Pit growth kinetics and conditions which support growth of a pit are not yet established fundamentally. Only empirical relationships exist that support stable pit growth subjected to critical ion concentration that must be maintained for pit growth as presented by Sato (1982). Such a model also is questionable since it requires a concentration gradient to develop due to diffusion. During hydrogen bubbling the mass transfer is increased several times higher than the critical rate required by Sato's model. Each of the above three mechanisms is not well accepted as a means by which a pit or pits should grow.

PITTING CORROSION FATIGUE MODEL

Hoeppner (1971, 1979) presented a fracture mechanics based pitting corrosion fatigue model in that pit depth was evaluated using pitting theory. Then, using linear elastic fracture mechanics concepts fatigue crack growth from the bottom of a pit was evaluated. This was an early attempt made to determine the transition from a pit to a crack. The model proposed by Hoeppner assumed hemispherical pits. This has been used by Kondo (1989) and Kondo and Wei (1989) in Newman and Raju criterion (1978) in calculating the stress intensity for a critical pit by assuming it to be a crack.

The pitting corrosion fatigue process has been conceptually separated into the following stages:

1. electrochemical stage and pit nucleation,
2. pit growth,
3. competative mechanisms in pit growth and fatigue crack nucleation,
4. chemically "short" crack growth,
5. "short crack" transition to "long crack",
6. long crack growth, and
7. corrosion fatigue crack growth and instability.

These stages are incorporated in the model presented in Fig. 1.

Figure 1 shows the pitting model using fracture mechanics concepts. It is clear from the model that external factors such as surface conditions, time of exposure, stress-strain conditions, environmental factors, thermal history and other factors decide the pitting nucleation and growth rates. Since pitting and its rate is time dependent, only after a critical time will the pits nucleate. Pit growth mechanisms not only depend upon the electrochemical parameters, but also, on several external factors as shown in the model. During this process two mechanisms namely; pit growth and fatigue damage by slip systems (clusters, slip saturation, intrusion or extrusion formation) compete with each other. The electrochemical growth rate of a pit is enhanced if the damage produced by the various processes interact with each other. Therefore, in a pitting corrosion fatigue process pit growth may transform into crack nucleation.

When the pit growth rate is smaller than the damage produced under fatigue for crack nucleation, the pit transforms to a crack. Therefore, in a latter stage when the crack formed grows, a "short crack grows chemically". However, mechanisms of crack growth in this stage have been investigated to a limited extent and such data needs generating. It is speculated that crack growth will occur in the directions of pre-damaged regions by corrosion, a crack may grow on the surface or sub-surface and not through the thickness crack

growth. Usual linear equations may not be useful to describe the crack growth behavior under such conditions.

Once propagation of a crack from the "short" region reaches a critical size, a long crack growth model as used by several workers be applicable to describe the corrosion fatigue crack growth behavior. However, it may be noted that selection of environmental and test parameters may produce accelerated crack growth or crack growth retardations. These aspects are very complex in nature and have not been accurately accounted for in computer codes (Gangloff et al, 1994). The four parameter Weibull equation may be used to model the corrosion fatigue crack growth rates as shown in Fig. 1.

FRETTING CORROSION FATIGUE MODEL

A similar model can be developed for fracture mechanics modeling of fretting fatigue. A fretting fatigue damage threshold concept was developed by Waterhouse (1972) and Hoeppner (1972) independently. It is realized that in a fretting corrosion fatigue situation the describing mechanisms of "short" crack growth will be a complex process since not only corrosion effects take place, but also, stresses produced on a surface due to relative normal load applications, all of these are difficult to model. Also, the frictional force effects must be accounted for in the model.

A limited number of fretting fatigue tests were conducted by Hoeppner (1976) and Cox (1979) for different materials in corrosive environments. The data are reported in Fig. 2 for a Ti-6Al-4V. It is interesting to note that fretting considerably reduced the fatigue life of Ti-6Al-4V. The corrosion fretting response of Ti-6Al-4V in different environments showed a marked reduction in life when the medium changed from air to distilled water and 3.5% NaCl solution as shown in Fig. 2. Fretting corrosion behavior was significant in 3.5% NaCl solution where the life reduction factor was 2 compared to air. A similar trend is shown in Fig. 3 for a 7075-T6 alloy.

The results presented above constitute some limited data in fretting corrosion fatigue. It also may be seen in Fig. 4 that fretting in corrosive environments produces pits and pit linking. As a result, a significant life reduction may occur under fretting situations when the contact stresses are produced on the protective film. The composition of the protective film varies within the layers. Therefore, properties of the film are likely to be uneven and the film may elongate at one place and rupture at another. Such local areas where there is a rupture in the film provide electrolyte entry. In addition, capillary action occurs and the electrolyte is spread within the metal under the film. Fretting conditions accelerate this process thereby, pits are produced earlier compared to in electrochemical conditions acting alone. An example of the fretting corrosion fatigue induced pit growth mechanism is shown in Fig. 4.

DISCUSSION ON PITTING CORROSION FATIGUE MODEL:

Effect of stress cycling under different environments such as lab-air, distilled water and 3.5% NaCl solution on the S-N type fatigue response of 7075-T6 was reported elsewhere (Hoeppner, 1971 and 1979, Ma and Hoeppner, 1994). Other data (Antolovitch and Saxena, 1985) showing this behavior were compiled and plotted collectively in (Ma and Hoeppner, 1994). The environmental effects were found to reduce the fatigue life significantly. Fatigue resistance of 7075-T6 aluminum alloy deteriorated more in 3.5% NaCl solution than in air. From the limited data that were available it was found that at fatigue lives of 10^6 cycles in air the maximum fatigue stresses reduced more than 3 times in 3.5% NaCl environment. In terms of cyclic life, the fatigue lives in air were an order of magnitude or more higher at the same stress levels from tests in 3.5% NaCl solution. Effect of frequency in a range of 5 to 20 Hz was observed to be very limited reported by many workers (Gangloff, 1988).

ELECTROCHEMICAL STAGE OR PIT NUCLEATION IN FATIGUE:

Under controlled environment fatigue tests, it is usually difficult to isolate the effects of interacting parameters such as the electrolyte, loads and those related to materials. Therefore, individual effects related to electrochemical aspects in corrosion and fatigue are difficult to separate if not impossible. In the nucleation stage under the conjoint action of cyclic loading and corrosion in a 3.5% NaCl solution, bubbling has been found as a feature. The rate of bubble generation depends upon time and temperature of exposure. Later, pits form at the regions of bubbling. There may be a dependency of bubble settlement on the surface and localized regions where the film is uneven or broken. Such sites are usually the sites where there are constituent particles present. Pit nucleation, though, depends upon the electrochemical parameters as mentioned in Table 1 and 2, microstructural features determine where a pit must nucleate. As a result, a detectable pit nucleation stage depends upon many factors which are not understood individually, in collective action their influence is very difficult to hypothesize at present. More research needs to be devoted to this area of research.

PIT GROWTH:

The kinetics of pit growth has been empirically modeled in terms of parabolic growth. Under the pitting corrosion fatigue process stable pit growth requires a critical ion concentration that must be maintained (Sato, 1982). However, bubbles as appear, transport hydrogen thus, the mass transfer rate is increased several times more than that required for stable pit growth (Sato, 1982). The effect of fatigue cycling and slip generation may provide electrolyte access in the metal surface and its entry is enhanced by capillary action. Therefore, these processes have not been evaluated quantitatively as affected by texture and microscopic features. In the case of several titanium alloys, hydride formation is an indication that the environmental effect is interacting with cyclic load response.

COMPETING MECHANISMS IN PIT GROWTH AND FATIGUE CRACK NUCLEATION:

Corrosion fatigue mechanisms of two or more independent processes acting conjointly under conditions where the combination is synergistic has not been investigated much.

In the literature, description of a hemispherical pit assumed as a crack has been reported and used as an ideal pit to model the pit as a crack. However, no attempts have been made to investigate the competition mechanisms between the electrochemical pit growth combined with the fatigue process which may nucleate cracks at multiple sites.

CHEMICALLY "SHORT" CRACK GROWTH:

At the bottom of a pit, depending upon microstructure these cracks grow intergranularly as found in the failure of many engineering components. These cracks may propagate faster than usual LEFM Mode I crack. Under corrosive environment, cyclic oxidation, embrittlement and local dissolution determine the crack growth rate, thereby, the crack growth is a chemical as well as a mechanical process. Damage by pitting corrosion and fatigue cracking, their multiple numbers and interactions among them is not known at the present time. Only limited tests that are available show that "short" crack growth is dependent on other parameters as well as dependent on the geometry of the specimen. A recent study by Piascik and Willard (1994) found intergranular crack growth in the "short" regime when the crack size in the z direction was less than 100μm and transgranular above it. In their study (Piascik and Willard, 1994) it was also found that at very low ranges of mode I ΔK (<1MPa√m) which was much below threshold mode I ΔK for long cracks in aluminum alloys (approximately 5 to 7 MPa√m) the crack growth rates increased. Thereby, consideration of threshold mode I stress intensity factor range of long cracks in the consideration of a pit becoming a crack is invalid. Recent work reported by Wei (1994) indicates that such a transition would occur at a mode I ΔK range of about 2.5MPa√m to 5MPa√m depending upon frequency and maximum stresses. Nearly same analogy was provided by Hoeppner (1971, 1979) over 20 years ago and is being studied further in author's laboratories. A new model shown in Fig. 1 is being proposed in this paper shows that subsequent to pit growth there is "short" corrosion fatigue crack growth which cannot be modeled accurately due mainly to limited data published and its behavior in corrosion fatigue is not investigated fully.

As undertaken in many studies (Hoeppner, 1971, 1979, Kondo, 1989, Kondo and Wei, 1989, and Wei, 1994) the criteria of pit transition to crack is incomplete as the role of short corrosion fatigue crack growth is ignored in all studies at very low stress amplitudes. Therefore, the transition stress intensity factor range derived by most authors is the transition where the short crack becomes a LEFM long crack and its growth can be modeled by linear equation presented by Paris. A co-operative program among various researchers working on this issue (Hoeppner, 1971, 1979, Kondo, 1989 and Kondo and Wei, 1989, Wei, 1994) will help expedite development of a method thereby, contribute to the understanding of various failures which propagated from the bottom of a pit and their sensitive usage parameters.

SHORT CRACK TRANSITION TO LONG CRACK:

As pointed out in this paper the "short" corrosion fatigue crack grows quite early in the pitting corrosion fatigue crack growth process within the range of mode I ΔK (<1MPa√m) at a rate much faster than the long cracks. Therefore, according to the models presented in (Hoeppner, 1971, 1979, Kondo, 1989 and Kondo and Wei, 1989, Wei, 1994) the transition is indeed not from a pit to a modelable crack but, transition from a "short" crack to a long crack.

Figure 5 shows a schematic plot of pitting corrosion fatigue crack growth process. The conventional models reported in (Hoeppner, 1971, 1979, Kondo, 1989, Kondo and Wei, 1989, and Wei, 1994) are represented by a thick line which has a zone that represents crack growth and mode I ΔK curve closer to threshold mode I ΔK or above. As a result "short" crack growth trend has not been incorporated. Should that behavior been incorporated, a range of mode I ΔK less than 1MPa√m to 2MPa √m would have shown crack growth rate (CGR) higher than the CGR of long cracks. Therefore, when a "short" crack growth rates become lower subjected to crack face reaching a hard particle or grain, the rate of CG becomes smaller and other factors such as time, corrosive environment, fatigue loads then determine its further growth or transition from "short" to long crack growth. As the range of CGR and mode I ΔK is enhanced above the threshold, the long crack grows. This has been widely examined in references (Hoeppner, 1971, 1979, Kondo, 1989, Kondo and Wei, 1989, Wei, 1994, Koch, 1994, and Brown and Srawley, 1967) as a pit transformation to a crack and corrosion fatigue crack growth.

LONG CRACK GROWTH:

In the data reported by Hoeppner, (1971, 1979), Kondo, (1989), Kondo and Wei, (1989), Wei, (1994), Koch, (1994), and Brown and Srawley, (1967) show that under pitting corrosion fatigue the range of linear behavior reduces. Depending upon frequency and environment, the crack growth rate behavior increases abruptly after a critical mode I ΔK. This value changes with materials and other parameters. For a 2024-T3 a shift from the steady crack growth to higher CGR occurred in a range of mode I ΔK from 24 to 28 MPa√m under pitting corrosion fatigue process (Koch, 1994). Such a behavior when the crack growth rate changes from point to point within the range of mode I ΔK range of long cracks, though observed under different environments reviewed in (Gangloff, 1988), has also been found for exposed materials of the C/KC 135 aircraft by Mills et al, (1995). Further studies of this behavior and corrosion fatigue crack growth rate modeling are presently underway in the author's laboratories.

CONCLUSION:

From this study the following conclusions are drawn:

1. The pit nucleation and growth stages in a corrosion fatigue process cannot be quantitatively separated. The number of fatigue cycles in a pitting corrosion situation to nucleate a "pit" or pits are considerably higher than pit growth and cycles to failure.

2. There are many pit growth mechanisms. However, in fatigue situations, it is very difficult to isolate electrochemical as well as mechanical effects. In such situations synergisms among various electrochemical and mechanical parameters determine the pit nucleation, growth, shapes and sizes and their numbers.

3. The pit growth due to electrochemical actions and crack nucleations are two competition mechanisms. As soon as the crack growth rate exceeds the pit growth rate, i.e., within the "short" crack regime, a pit becomes a crack.

4. A model presented in this paper of pitting corrosion fatigue and fretting corrosion fatigue has seven stages. Each one of these stages is not very well understood. The seven stages are as follows:
   1. electrochemical stage or pit nucleation in fatigue,
   2. pit growth,
   3. competition mechanisms in pit growth and fatigue crack nucleation,
   4. chemical "short" crack growth,
   5. short crack transition to long crack,
   6. long crack growth, and
   7. corrosion fatigue crack growth modeling.

5. Most models presented considered the transition of "short" crack growth to "long" crack, thereby, application of LEFM mode I ΔK can be applied. It is proposed in this paper that from the bottom of a pit "short" crack grows and the critical parameters related to pit depth and beginning of "short" crack growth needs defining by appropriate experimental procedures.

6. More work is required to develop a methodology of a pit transition to a crack and the proposed seven stages of pitting corrosion fatigue process so that it can be computed in structural integrity assessment tools and future usage parameters be estimated for various components.

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Zahavi, J. and Yahalom, J., 1975, in proc. of 6th International Congress on Met. Corrosion, Sydney, paper 9-9.

Table 1. Summary of pit nucleation mechanisms.

Adsorption related	Ion migration	Film breakdown
Based on adsorption of aggressive anions at energetically preferred sites. Only above a critical potential, Cl adsorption takes place by breakdown of the localized passivity.	Penetration of anions from oxide/electrolyte interface to the metal/oxide interface or migration of cations as a decisive process. Only after a critical breakdown potential is reached the penetration occurs. Rapid cation egress results in pitting.	Independent as well as a interdependent phenomena of the other two mechanisms. Breakdown of a passive film provides electrolyte direct access to the metal surface and pitting occurs.

Table 2. Summary of pit growth mechanisms.

Charge transfer based	Diffusion related	Resistance controlled
At the bottom of a pit a differential current density exists that causes a pit to grow. A constant current is assumed as it is very difficult to measure the current and its profiles with respect to the depth within the pit.	Salt powders present on the surface diffuse by causing a breakdown of the protective film in aluminum alloys.	Pitting also occurs at a high Ohmic-limited current density. Hydrogen bubbles generated within a pit increases the mass transport rate. Pit growth rate is a time dependent as well as applied potential dependent process. Increasing the potential enhances the rate. A parabolic growth often is observed.

SEVEN STAGES OF PITTING CORROSION FATIGUE MODELING

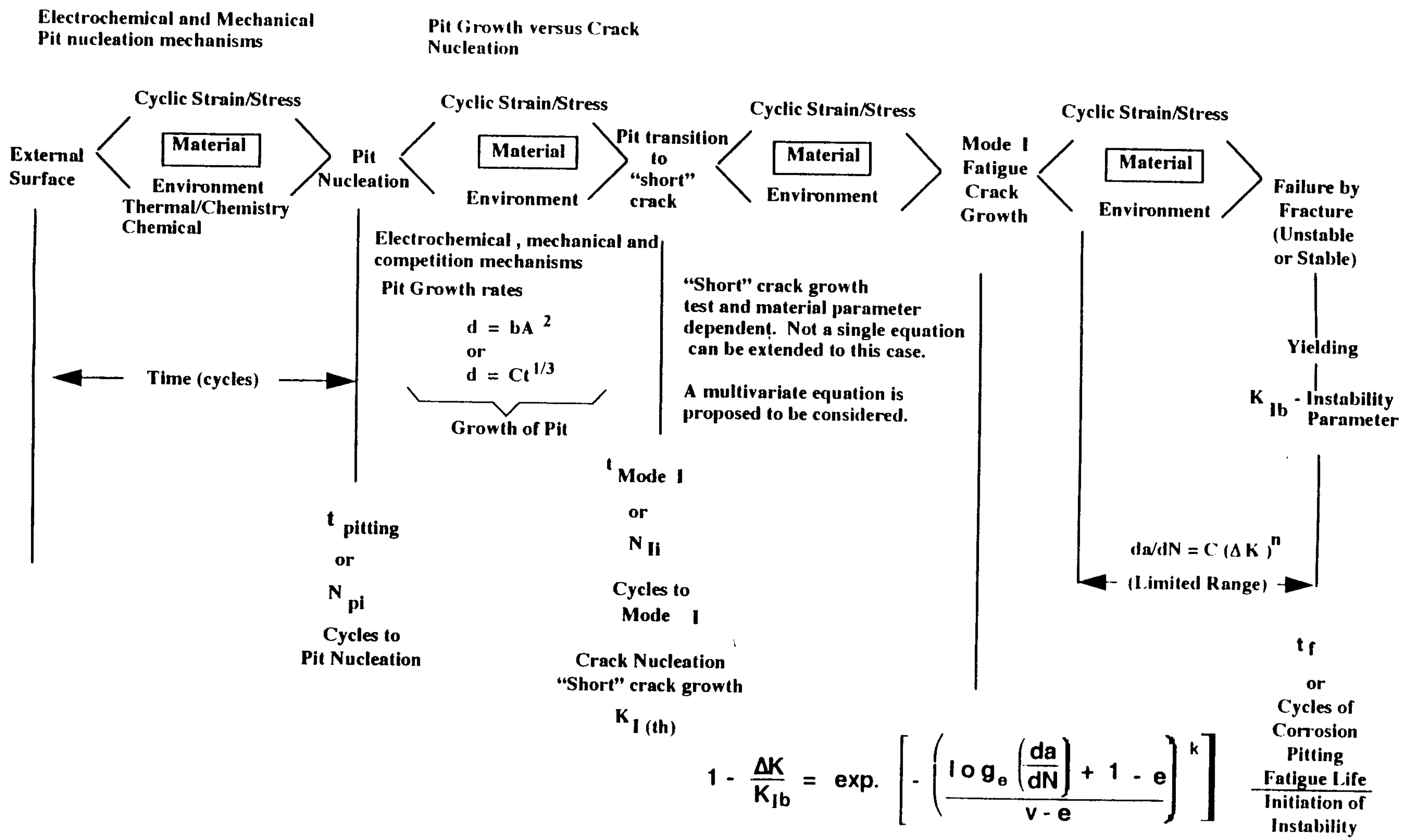

Fig. 1. Seven stages of pitting corrosion fatigue model.

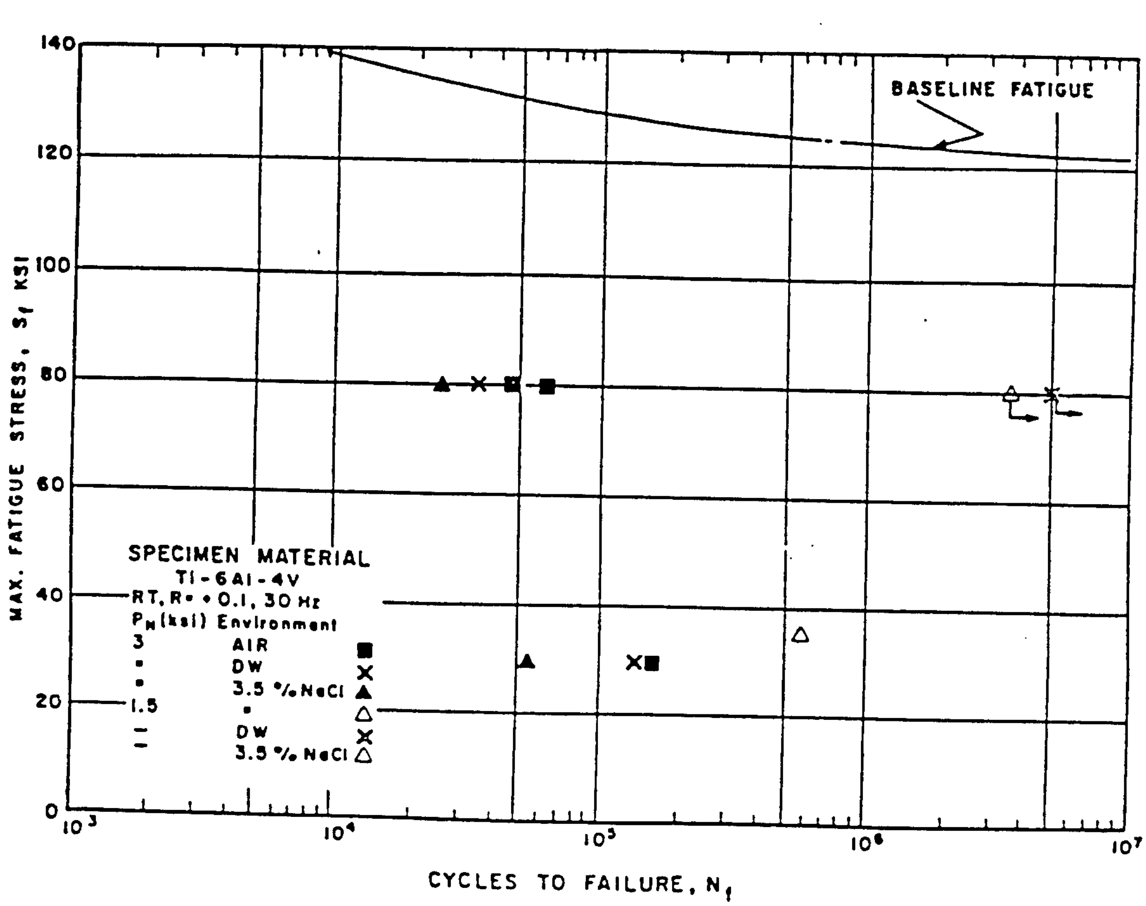

Figure 2. Baseline fatigue and corrosion fretting fatigue behavior of Ti-6Al-4V.

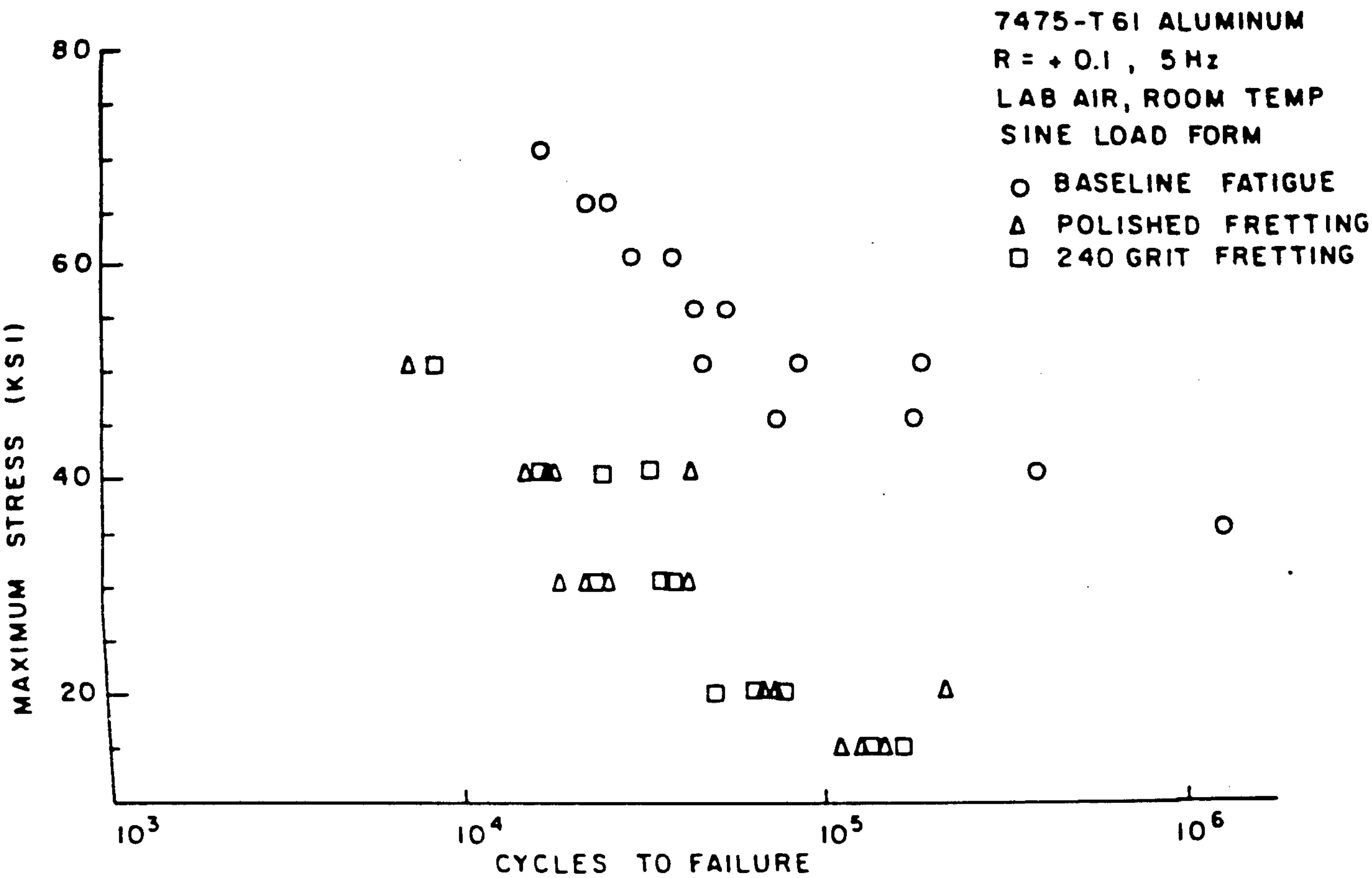

Figure 3. Baseline fatigue and corrosion fretting fatigue behavior of 7075-T6 aluminum alloy.

Figure 4. Fretting induced pitting of 7075-T6 aluminum alloy in 3.5% NaCl solution

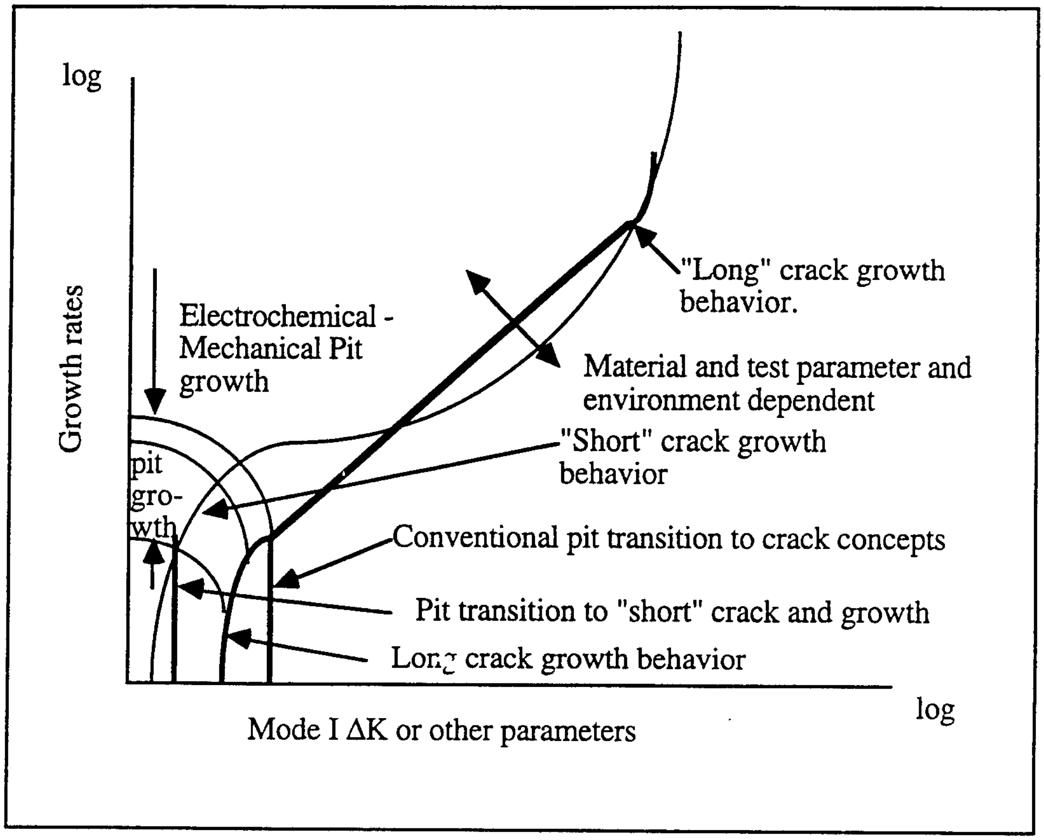

Fig. 5. Schematic representation of pit transition to crack concepts.

Literature Search and Review of "Short" Crack Behavior of Structural Materials -- PART I - Microstructural and Environmental Aspects

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Submitted as a part of the research program
"Investigation of Modeling the Fretting and Pitting Corrosion Fatigue Processes In Structural Aluminum Alloys"

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EXECUTIVE SUMMARY

The Quality and Integrity Design Engineering Center (QIDEC), Mechanical Engineering Department at the University of Utah conducted extensive literature search and review on the effects of environment and microstructure on the behavior of "short" cracks in structural materials. This work was performed under the Airforce Office of Scientific Research (AFOSR) grant as a part of the research program entitled "Investigation of Modeling the Fretting and Pitting Corrosion Fatigue Processes In Structural Aluminum Alloys". The findings from the literature search and review are presented herein as an interim report (PART-I) for (AFOSR).

As the meaning of the word "short" as related to the formation and arly propagation of cracks differs significantly with respect to the conditions of interest (coi), it has been realized by the technical community that there are three basic types of "short"/"small" cracks viz. microstructurally-short-cracks, chemically-short-cracks, and mechanically-short-cracks. The most important finding of the literature search process showed that while a tremendous amount of effort has been expended during the past several years to characterize the "short" crack behavior of structural materials, very little research has been performed as related to environmental issues such as the contribution of chemical environment to the formation and early propagation of cracks as well as "short" cracks as related to the pitting corrosion fatigue mechanism. Moreover, the literature search and review process has shown that the technical community has not done any work to characterize the "short"/"small" crack behavior of materials under the synergistic process of fretting corrosion fatigue.

This review report addresses the microstructural and environmental "short" crack studies that have been conducted since 1965. Section 1 of this report gives an introduction summarizing the present understanding of the mechanisms of "short" crack behavior of materials. This is followed by a detailed review of

microstructurally "short" crack studies including the effects of grain size, grain and phase boundaries, variation of precipitate size, processing technique, and stress ratio. In addition, "short" crack studies as related to stage I crack propagation also are discussed in section 2. Section 3 discusses environmental influences in the "short" crack regime. Conclusions are given in section 4. Some "short" crack test results are extracted from the literature and they are included in appendix I. Experimental details including the commonly used "short" crack specimen geometry, test techniques and test parameters are tabulated in appendix II. A global view of the "short" crack challenge in materials is schematically presented in appendix III. Also, a title list of "short" crack related works published so far is given in appendix IV.

It is envisioned that Part II of the report will address some preliminary experimental investigations on "short" crack fretting fatigue and pitting corrosion fatigue studies in 7075 aluminum alloy specimens. In addition, in this future report, "short" crack modeling methods that are published in the literature will be discussed and also these modeling methods will be compared with the findings from the proposed "short" crack fretting fatigue and corrosion experimental studies. Two to three "short" crack fretting fatigue tests and some pitting corrosion fatigue "short" crack tests will be performed using replication technique. Part II of the report incorporating preliminary test data will be submitted to AFOSR after the completion of the testing program.

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LIST OF FIGURES (Provided in Appendix I)

Figure 25. Literature results demonstrating the effect of crack size on corrosion fatigue in steels exposed to aqueous environments (Gangloff and Wei, 1986)

1. Introduction

Design of structural components using fracture mechanics concepts requires three basic parameters viz. load or applied stress, stress intensity factor, and discontinuity or crack size. Conventionally damage tolerant design methods consider an initial "flaw" size typically in the order of 1 mm (0.04") while applying fracture mechanics concepts to design damage tolerant components (Potter and Yee, 1983). Fatigue crack growth tests in the "long" crack regime as per ASTM E647 are conducted on materials that are to be used to make the damage tolerant parts and together with the assumed initial "flaw" size as well as the appropriate loading and material parameters inspection intervals are computed. However, the technical community has realized the significance of the formation and existence of the so called "short" cracks (examples: manufacturing discontinuities like crack(s) from a hole that may result from improper riveting operation, and material microstructural discontinuities in the "short" crack range, typically in the order of grain size, or less) and their growth is "faster" when compared to "long" cracks in the equivalent or even at lower stress intensity range. With the support of experimental studies conducted by several researchers to determine the "short" crack behavior of materials, it can be argued that the current practice of computing inspection intervals of aircraft critical structural parts using the "long" crack growth data and the initial "flaw" size of 1 mm might lead to unexpected fatigue crack growth behavior resulting in a catastrophic failure. This implies that designing components using

damage tolerant concepts may still not be safe when "short" crack behavior is not integrated in fatigue life prediction methods. "Small" cracks were found in aircraft fuselage riveted lap joints (Schijve, 1992), fastener holes (Potter and Yee, 1983) and in tear-down aircraft wing critical structures (Wood and Rudd, 1983).

Although many research studies were conducted, there is still not a clear understanding among the researchers with regard to the usage of terminology such as "small" and "short" as related to the size of the crack as these terms are often mixed up in the literature and this alone exemplifies the complexity of the challenge we have at present. However, many "short" crack researchers agree on the following aspects:

• There are three basic types of "short" cracks viz. mechanically or physically "short", microstructurally "short" and chemically "short" (McClung, Chan, Hudak and Davidson, 1994).
• "Short" cracks can not be modeled using the Linear Elastic Fracture Mechanics (LEFM) concepts although some workers have attempted to convert the "short" crack data to "long" crack, to compare the crack growth behavior in both the regimes and also to evaluate the "effective crack driving force" for "small" and "large" fatigue cracks (Tokaji, Ogawa, Harada and Ando, 1986; Tokaji, Ogawa and Harada, 1987; Davidson, 1988; Hyspecky and Stanadel, 1992; Nicholls and Martin, 1991; Sheldon, Cook, Jones and Lankford, 1981).
• "Short" cracks often grow "rapidly" when compared to "long" cracks at a lower stress intensity range that is below the "long" crack

threshold and also at an equivalent stress intensity range. In some materials, the "short" growth rate is observed to be much "faster" than would be predicted by extrapolating "large" crack data (Lankford, 1982; Lankford, 1985).

• The material parameters which influence plastic deformation viz. grain size, grain orientation, texture, work hardening rate, slip band character, local microscopic fracture toughness, inclusion size and content as well as second phase particles have an important role in "short" crack growth (Miller, 1982).

• Current NDI technologies are not capable of detecting the discontinuities that are in the "short" crack range (Wood and Rudd, 1983). This directly affects the inspection related issue as it is inextricably linked with the damage tolerant design concepts.

• There is no single parameter that can define the "short crack driving force".

• Although a few studies have attempted deterministic "short" crack growth prediction methods for physically or mechanically "short" cracks, as the influence of the microstructural variations on the "short" crack behavior of materials is extremely complex, the challenge of incorporating "short" crack methodology into present practice of fatigue life prediction analyses must be treated with probabilistic approaches. This necessitates the study of scatter in the behavior of "small" cracks to understand the physical basis of scatter in the fatigue lives of components or specimens (Goto, 1992 and Goto, 1993).

Several researchers have postulated different mechanisms for the behavior of "short" cracks. They are indeed related to the type of

"short" cracks. For microstructurally "short" cracks, "crack tip shielding" and "enhanced crack tip plastic strains" are stated to be responsible phenomena (Ritchie and Lankford, 1986). Also, evidences suggest that microstructurally "short" cracks are sometimes obstructed locally by grain boundaries (crack arrest), influenced by non-uniform growth and sometimes may experience higher cyclic plastic strains at the crack tips resulting in "faster" growth (Lankford, 1981, Lankford, 1982, Ritchie and Suresh, 1995). Moreover, "short" cracks may be subjected to "crack deflection" (Suresh, 1983) that may be related to the orientation of each grain. Thus, texture has an important role in determining the behavior of microstructurally "short" cracks.

Excessive plasticity is stated to be the mechanism for mechanically "short" cracks. This implies that the assumption of "small-scale yield" that forms the basis of linear elastic fracture mechanics is not applicable. Local crack tip environment (Gangloff, 1985) has been hypothesized as a predominant mechanism for the "faster" propagation of chemically "short" cracks as they are more vulnerable to chemical attack than "long" cracks because of the relative ease of access to the crack tip. For physically "short" cracks, it is hypothesized that the "crack closure effect" that decreases the "crack driving force" for "long" crack propagation may be absent (Schijve, 1986). Furthermore, it is believed that "short" cracks that are usually associated with a limited "wake" are less able to develop the same magnitude of shielding as equivalent "long" cracks at the same nominal stress intensity range (Lankford 1986). Some

experimental studies have showed that although the growth rate of microstructurally "short" cracks is "faster" initially, once the crack tip reaches the grain boundary the growth rate either gets reduced or some times completely arrested at the grain boundary (Lankford, 1982, Lankford, 1983, Lankford and Davidson, 1986, Tokaji and Ogawa, 1992, Ritchie and Suresh, 1995).

Moreover, the dimension of the crack also is related to the types of crack. Mechanically "short" crack size 'a' is considered to be less than the plastic zone size and microstructurally "short" crack size is related to the grain size (Ritchie and Suresh, 1995). As the crack length is "short" compared to the microstructural dimensions, such as the grain size, it has been realized that the assumption that the crack grows in a homogeneous, isotropic continuum is no longer valid.

Therefore, the primary focus of this report is to review the present understanding of the influence of microstructure and environment on the behavior of "short"/"small" cracks in structural materials.

2. The effect of microstructure on the "short"/"small" Crack behavior of materials

This section reviews microstructurally "short"/"small" crack studies conducted on materials. The mechanisms of the "short" crack growth behavior as related to the local microstructural variations and crack closure phenomena are discussed in detail. Some "short" crack test results are reproduced from literature and they are included in appendix I. The experimental details including specimen geometry as well as test techniques corresponding to the following discussion are included in appendix II.

When applying damage tolerance principles to the design of structural components subjected to cyclic loading, a geometry independent parameter ΔK (stress intensity range), is used to characterize the crack growth. This is the basis of Linear Elastic Fracture Mechanics (LEFM). Moreover, it is presumed that cracks would not propagate at a stress intensity below a threshold value and it is usually denoted as ΔK_{th}. Although it has been realized that there is no geometry independent crack driving force in the "short" crack regime, several attempts have been made to predict "short" crack growth rates including some approaches that are based on crack deflection, crack closure, 'J' integral and some semi-empirical methods (Morris and Buck, 1977, McEvily and Minakawa, 1984, Suresh, 1985, and El Haddad, Dowling, Topper, and Smith, 1980). As microstructure has varying effects in materials, it has been realized that the deterministic way of fatigue life prediction in the "short"

crack regime is far from accurate. However, some researchers have proposed models which incorporate microstructural parameters into the "short" crack growth rate equations (Petit and Zeghloul, 1986, Petit and Zeghloul, 1990)

The factors that are believed to be responsible for the "faster" growth of "small" cracks as summarized by Schijve are:

- The front of a "microcrack" is more regular than the front associated with a "long" crack.
- A single slip system is required for propagating a "small" crack whereas several systems are necessary for "macrocracks".
- Anisotropy effects, grain boundary structure and inclusion content may influence "small" crack behavior.
- The reversed plastic zone behind the crack tip may be different and may induce "crack closure effects" which are a function of the crack length.
- The roughness of the fracture surfaces may play an important role in favor of the crack closure phenomenon.

A distinction between "small" and "short" crack growth behavior was made by some researchers (Lankford and Davidson, 1986, Breat, Mudry, and Pineau, 1983) by experimental studies. Their studies have showed that the experimentally measured difference in crack closure response between "short" and "large" cracks in A508 steel may be sufficient to explain differences in their crack propagation behavior. From experimental studies conducted in A508 steel, it was observed that da/dN versus ΔK for "short" cracks

was only an extension of the "long" crack behavior into the subthreshold regime. This is illustrated in Fig. 1 (Breat, Mudry, and Pineau, 1983). As can be seen in Fig. 1, "short" and "long" cracks exhibit the same behavior. However, "small" cracks are related to a controlling microstructural element, usually the grain size, and these cracks are observed to be distinct from "short" cracks, i.e., through thickness cracks (0.5-2.0 mm) in length. (Taylor and Knott, 1981) have suggested that the "small" crack regime corresponds roughly to microcracks such that 2a = 10 GS where GS is grain size. Also, (Breat, Mudry, and Pineau, 1983) have showed that "short" cracks do not grow much faster at a given ΔK than do "long" cracks. But "short" cracks have a lower threshold, above which they follow roughly the same da/dN versus ΔK curve that a "long" crack would if it were simply to continue growing below its threshold (see Fig. 1). However, "small" cracks, as hypothesized by these researchers, not only grow below ΔK_{th}, they also grow much faster than would be predicted by extrapolating long crack results below their threshold. Some studies (Taylor and Knott, 1981, Hicks and Brown, 1984, Lankford and Davidson, 1986) have showed that "small" crack growth converges with the "large" crack curve when plastic zone size is approximately equal to the grain size. This phenomenon has been observed in aluminum alloy, fine grained, coarse grained and single crystal astroloy, and p/m aluminum alloy as shown in Figs. 2, 3, and 4.

In many cases it has been shown that the arrest or retardation of small cracks correlates with the crossing of grain boundaries (Lankford, 1982, Lankford 1985). It also has been postulated that if

the plastic zone size is less than the grain size such crossings are infrequent. In comparison with the "large" cracks, cracks are observed to grow in many unfavorable grains simultaneously, hence, the average rate of growth is much lower than that for "small" cracks at equivalent ΔK levels in favorably oriented grains. Although these studies attempted to distinguish "small" and "short" cracks, many of the researchers used the terminology in a mixed manner. Therefore, as the purpose of this report is only to review the present understanding with regard to the fundamental issues pertaining to microstructural and environmental aspects of the fatigue crack formation and growth, this report includes the terms "short" and "small" as they are extracted from the literature.

In general, the "faster" growth of microstructurally "small" fatigue cracks has been observed to be associated with second phase particles (Pearson, 1975), inclusion particle clusters or voids (Newman and Edwards, 1988), eutectic colony boundaries (Taylor and Knott, 1981), and grain boundaries (Lankford, 1985).

Early in 1975, bending fatigue studies were performed on commercial aluminum alloys viz. aluminum-copper-magnesium (BS L65) and DTD 5050 (aluminum-zinc-magnesium) (Pearson, 1975). It was observed that cracks in the order of, or even less than, the grain size grew "faster" than "long" cracks. i.e. the mean crack growth rate in the early stage was observed to be 1.27 X 10^{-6} mm (5 X 10^{-8} in.). In this study crack formed at surface inclusions and it was related to previous cold working that was performed on the material. Fracture

mechanics approach was used to calculate ΔK values for "short" and "long" cracks because the plastic zone size was observed to be 1/20th of the crack length (see Figs. 5(a) - 5(d)). Also, it was concluded that the growth rates in the early stage was much "faster" than would be predicted from the "long" crack data. The results of experiments performed by Pearson for alloy BS L65 and DTD 5050 are given in Table 1 and 2 and the plots of the number of cycles to "initiate" a crack of 0.05 mm (0.002 in) for the two alloys are given in Figs. 6 and 7.

Following this study, in 1976, uniaxial fatigue loads were applied on Aluminum alloy 2219-T851 parallel to the rolling direction in room temperature and it was observed that crack nucleated at the surface at intermetallic inclusions (Morris, Buck and Marcus, 1976). Moreover, it was showed that at the fatigue loads that are less than 0.6 times the yield stress several cracks coalesced to form a macrocrack that lead to the ultimate failure. The most important finding of this study showed that there was significant retardation of microcracks with grain boundaries. Similar observation was made in another study (Lankford, Cook and Sheldon, 1981), in that it was hypothesized that "small" cracks grew "rapidly" during the initial stage and when the crack tip interacted with grain boundaries, the growth rate slowed down. However, it also was formulated that as the crack moved away from the boundary, the crack growth became "faster" again. This behavior was related to local microplasticity in certain preferentially oriented grains (Lankford, 1982). This study was conducted in laboratory air in

7075-T6 (precipitation hardened aluminum) with 60% humidity. It was observed that for the "smallest" crack (2a<40 μm), da/dN did not increase monotonically with ΔK. However, there was a decrease in the crack growth rate and reached a minimum in the range 30 μm < 2a < 40 μm. Some of the cracks were found to be "nonpropagating". Lankford also proposed a schematic representation of the "short" and "long" crack growth behavior as shown in Fig. 8. It was concluded that "small" cracks could become "large" cracks *"when their LEFM plastic zones begin to exceed in size the maximum grain dimension"*.

Similar observation was made when "small" crack behavior was studied in A286 steel in which the "short" crack effect disappeared when the crack-tip plastic zone size became greater than the grain size (Mei and Morris, 1993). This study also supported the hypothesis that peak stress and microstructural effects in addition to the absence of crack closure are some of the factors that influence the "short" crack growth in this material. In another study, it was observed that when the applied stress was sufficiently high the "short" crack growth rate could be sustained and could overcome the microstructural barrier when tested in plain specimens (Pan, De Los Rios and Miller, 1993). Also, in this study, tests on notched specimens (8090 Al-Li alloy) revealed that the extent of notch tip and crack tip plastic zones control "short" crack propagation. It was hypothesized that *"a short crack will continue to propagate only if its own plastic zone can sustain growth as the crack tip extends beyond the notch zone"*. This study presented some interesting data on the "short" crack growth from corner notches as shown in Fig. 9. Also, the

effects of grain boundary, viz, changing the direction of the crack, temporary stopping of a crack, and forcing the crack to adopt a zig-zag path were observed.

In SiC reinforced aluminum alloy (6061), "short" cracks were found to propagate through the SiC particles as the crack front fractured the particles (Kumai, King and Knott, 1990). In another study, the presence of voids in an overaged 2024 aluminum alloy was observed to be responsible for the formation of fatigue "micro cracks" (Sigler, Montpetit and Haworth, 1983). In this study, the density of microcracks (cracks were counted when the size was more than 5 μm) was observed at different stress amplitudes during the fatigue life. It was found to be 300/mm^2 at 320 MPa, in 771 cycles at failure, but less than 0.5/mm^2 at 200 MPa, in 1.5X10^5 cycles (at failure). This result has very high significance as related to the possibility of microcracks coalescence into macrocracks that may affect the ultimate fatigue life.

In nodular cast iron, "short" cracks formed from either the graphite nodules or microshrinkage pores (Clement, Angeli and Pineau, 1984). Crack closure effect was suggested as the mechanism for the "faster" growth of the "short" cracks for the given stress intensity factor. On the other hand, in a medium carbon steel, the "short" crack growth rate was related to the intensity and the extent of plasticity of the crack tip (De Los Rios, Tang and Miller, 1984). Also, it was observed that "short" crack growth decreased or even arrested at ferrite-pearlite boundaries. However, as the stress level

was increased to certain value, the two of the arrested cracks at the ends of the ferrite joined up and the resultant crack branched off along the prior austenite grain boundaries. Therefore, it was suggested that the critical fracture occurred when two of the "short" cracks joined and branched that made it possible to propagate into the pearlite. Similar hypothesis can be postulated for a "short" crack nucleating from corrosion pit in aluminum alloys and the stress concentration of the pit may be sufficient for "short" crack to overcome the grain boundary barrier for subsequent propagation. It can be further hypothesized that if cracks form from adjacent pits these "short" cracks can join and may grow "faster" when compared to "microcracks" coalescence resulting from other microstructural heterogeneities. However, there are no experimental data to support this theory and the very possibility of this occurring in a material should be investigated.

"Small" surface crack (2 to 1000 μm) studies on other materials like aluminum-lithium alloy 2090-T8E41 also have showed an accelerated growth at ΔK levels as low as 0.7 MPa $\sqrt{m}$ (Venkateswara Rao, Yu and Ritchie, 1988) and this behavior was related to restrictions in the development of "crack tip shielding" resulting from "roughness-induced crack closure". The test results are shown in Figs. 10(a) and 10(b).

In low carbon steel, when fatigue tested with specimens produced with two ferrite grain sizes of 24 and 84 μm, it was found that in fine grained material most of the cracks formed within ferrite

grains and in coarse grained material the cracks formed at grain boundaries as shown in Fig. 11 (Tokaji, Ogawa and Harada, 1986). The most important finding of this study was that the effect of grain boundary strongly depended on "short" crack length. For fine and coarse grained material, when 2c>200 μm and 2c>250 μm respectively, the grain boundary was not a barrier for the "short" crack propagation and dc/dN increased with increasing crack length (see Fig. 12). Fig. 13 shows the relationship between crack growth rate and stress intensity range for "small" fatigue cracks. Also, other studies showed that the effect of a grain boundary on the "short" crack growth appeared to be dependent on the orientation of ferrite grains (Lankford, 1985 and Suresh and Ritchie, 1984). In general, it has been recognized that when cracks are of a length comparable to the scale of the microstructure, the growth is greatly affected by the microstructure and the relevance of continuum mechanics is limited. The concept of "microstructural dissimilitude" was proposed to explain this kind of behavior (Chan and Lankford, 1988).

During the growth of "short" cracks, if the crack front behaves in a similar fashion when it intersects many grains irrespective of the crack length, these cracks are believed to possess similitude and the stress intensity range (corrected for yielding) can be used to correlate the crack growth rates. However, for "short" cracks, it has been observed that there is no linear relation between da/dN and ΔK. Thus, the challenge becomes so complex as it is highly impossible to predict the potential site of crack nucleation that may be related to the preferred orientation of the grain for the cracks to form. More

importantly, because of the greater dependence of the "short" crack challenge on the microstructural variations in a material that is frequently in the order of a grain size, the "short" crack problem is intrinsically statistical in nature. Therefore, to improve the resistance of the material microstructure to the nucleation of "short" fatigue cracks, alloy design incorporating "small" randomly oriented grains, and texture to have only a few grains as possible for an easy crack growth under a known loading condition was suggested to be a useful way to deal with this issue (Ritchie and Lankford 1984). Also, the challenge becomes still more complex as the local fracture toughness of each grain is of practical significance as related to the formation and the propagation of cracks in the "short" regime.

The effect of processing techniques on the "short" crack behavior was studied in an aluminum-magnesium-silicon alloy (Plumtree and O'Connor, 1991) in strain control. "Short" cracks were observed to form from second phase particles and the growth was impeded at grain boundaries. It was concluded that the extruded alloy with a finer microstructure and smaller second phase particles demonstrated a superior resistance to formation of "short" cracks when compared to squeeze-cast material. "Short" cracks in the order of 3 to 147 μm were found to form from Ti_3Al hcp alpha phase in a titanium aluminide alloy (Davidson, Cambell and Page, 1991). The results showed that the growth of "small" cracks in titanium aluminide alloy was slower by a factor of 10 to 100 when compared to aluminum alloys.

As the grain size has an important role in the behavior of microstructurally "short" cracks, a study was conducted to demonstrate the effect of grain size on the "short" fatigue cracks. Specimens from 7075-T6 were prepared to produce grain sizes of 12 and 130 μm. Growth rates of surface cracks in the order of 20 to 500 μm were studied during the axial fatigue test in laboratory air of 45 to 60% RH (Zurek, James and Morris, 1982). It was found that the mechanism for the "short" crack growth behavior was dependent on alloy grain size.

As discussed so far, many of the published works on the behavior of "short" cracks indicate that grain boundaries impede propagation, resulting in decrease in the growth rate or complete arrest in some cases. This kind of behavior has been modeled by some workers (Eastabrook, 1984, Hobson, 1982, Lankford, 1982). These models predict that increasing grain size will lead to faster crack propagation rates in the "short" crack regime. A schematic to represent the predicted effect of grain size on short crack growth is given in Fig. 14 (Lankford, 1982). (Brown, King, and Hicks, 1984) supported this model when they conducted "short" crack studies on a Ni-base superalloy, Astroloy, with grain sizes of 12 and 50 micrometer. They observed slower crack propagation rates in the fine grained material than those in the coarse grain size. Fig. 15 shows the effect of grain size on the fatigue crack growth rates in Astraloy resulted from this study. Also, (Wagner, Gregory, Gysler, and Lutjering, 1986) showed a similar trend in Ti-8.6Al alloy with grain sizes of 20 and 100 μm. In addition, some studies (Hirose and

Fine, 1983) reported slower growth in a powder metallurgy aluminum alloy with a fine grain size. However, a different behavior was observed (Taira, Tanaka, and Hoshina, 1979) in a 0.2% C steel with a ferrite/pearlite microstructure. They observed similar growth rates at ferrite grain sizes of 20.5 and 55 μm and lower growth rates only when the grain size was reduced to 7.8 μm. This is illustrated in Fig. 16. In contrast to these results are those of (Brown and Taylor, 1984, and Zurek, James and Morris, 1982). Their studies (Brown and Taylor, 1984) in a mill annealed alpha/beta titanium alloy (Ti-6Al-4V) with grain sizes of 4.7 and 11.7 μm, could not detect any grain size effect. Also, (Zurek, James and Morris, 1982) in 7075-T6 aluminum alloy, observed a decrease in growth rate with increasing grain size from 12 to 130 μm. Fig. 17 illustrates the results from this study. Thus much of the published work on the effect of grain size on "small" fatigue crack growth indicates that increasing grain size leads to "faster" crack growth rates. In the microstructurally "small" crack regime, this trend is related to the difference in the "blocking effect" of grain boundaries. This so called "blocking effect" was observed to occur more frequently in fine grained material than in coarse grained material. As demonstrated (Tokaji, Ogawa, Harada and Ando, 1986) in quenched and tempered steel, prior austenite grain boundaries act as barriers to the growth of microstructurally "short" cracks. It was observed in fine and coarse grained materials as shown in Fig. 12.

The effect of variation of precipitate sizes on the "short" crack behavior was studied (Brown, King and Hicks, 1984) using Astroloy

with 50 μm grain size. In this study, they investigated a range of gamma prime distributions, achieved by different heat-treatment processes and they found similar "short" crack growth rates in all conditions. Also, similar observation was made in a nickel base super alloy, Waspaloy, with about the same grain size, but somewhat lower gamma prime volume fraction. The results from this study are shown in Fig. 18. It also was realized by some investigators that one of the important ways of controlling short crack behavior in steels and titanium alloys was through variations in the distribution and proportions of the phases present. (De los Rios, Tang and Miller, 1984) When tested in a 0.4% C steel with a strongly banded structure with alternate layers of alpha and pearlite, observed cracks nucleating in the alpha and were held up by the alpha/pearlite interfaces and sometimes the propagation was completely stopped after reaching these boundaries. Also, in another study using the same material (De los Rios, Mohamed and Miller, 1985), but with a different microstructure in which the alpha outlined the prior austenite (gamma) grain boundaries, the similar observation was made.

In 0.37% C steel (Hoshide, Yamada, Fujimura, 1985), with an alpha/pearlite microstructure in two conditions, air-cooled from 865°C and furnace-cooled from 940°C, cracks formed in the alpha/pearlite boundaries and stopped when the cracks reached the pearlite. Also, slower "short" crack growth rates were presented in the air-cooled material with the finer distribution of alpha. In one of the earlier studies (Kunio and Yamada, 1979) using martensite steel

(alpha prime/ferrite alpha mixtures consisting of about 50 volume % of each), crack formation occurred in the alpha and growth stopped on reaching alpha prime. However, it was found that at higher stress levels these cracks continued to propagate, but until they reached lengths of 400-500 μm, their growth was still impeded by regions of alpha prime. From these studies, in steels, it was clearly observed that the presence of harder phases forced the cracks to take a tortuous path by deflecting the crack paths. Similar study was carried out in titanium alloys (Hicks and Brown, 1984). They compared the behavior of a beta processed titanium alloy, IMI 685, with a coarse aligned alpha structure with that of alpha/beta heat-treated IMI 318. The changes in orientation of the alpha plates at the prior beta grain boundaries and the coarse beta grain sizes produced at the high beta heat-treated proved to be the main barriers to "short" crack propagation in the IMI 685. However, in IMI 318 which consisted of regions of primary alpha and transformed beta, cracks formed in the alpha and were impeded by the harder regions of transformed beta, thereby meeting the effective barriers at lengths very much shorter than the prior beta grain size. As shown in Fig.19, a five-fold difference in average growth rate was obtained between the two microstructures.

Also, another study in IMI 318 and IMI 550 with different heat-treatment processes (Boilngbroke and King, 1986) clearly showed that a finer and harder transformed beta impedes the crack growth more effectively than a coarser transformed beta produced at a slower cooling rate. Moreover, they showed that an average

"short" crack growth rates in the beta heat-treated structure are up to an order of magnitude faster than in an alpha/beta heat-treated condition in IMI 318. Similar observations were made in dual phase steels with ferrite and martensite phases, crack growth rates and crack path were strongly affected by martensite phase (Minakawa, Matsuo, and Mcevily, 1982, Dutta, Suresh, and Ritchie, 1984, and Shang, Tzou, and Ritchie, 1987). Somewhat similar research observations also were presented (Tokaji and Ogawa, 1988) in medium carbon steel and dual-phase stainless steel as shown in Fig. 20). As can be seen from this illustration, it was found that as cracks grew into pearlite phase from ferrite phase in medium carbon steel and austenite phase from ferrite phase in dual-phase stainless steel, crack growth rates showed a marked decrease at phase interfaces. Also, it was observed that cracks tend to grow predominantly within ferrite phase in medium carbon steel. Fatigue crack was also found to form in softer ferrite in C-Mn steel (de los Rios, Navarro, and Hussain, 1992). Moreover, (Kawachi, Yamada, and Kunio, 1992) showed the possibility of coalescence of "small" cracks in a dual-phase (martensitic-ferritic) carbon steel and as a result of this, cracks increased their length along the matrix-ferrite, by-passing the harder martensite. They also demonstrated that the crack coalescence could be suppressed in this kind of steel by preparing a dual phase microstructure with the matrix ferrite enclosed by the second phase martensite. This resulted in increase in the fatigue strength of this steel.

Also, other microstructurally "short" crack studies in different metals viz. low carbon steel (Tokaji, Ogawa, and Harada, 1986), medium carbon steel (Tokaji and Ogawa, 1988), high tensile steel (Tokaji, Ogawa, and Harada, 1987), low alloy steel (Tokaji, Ogawa, Harada and Ando, 1986), aluminum alloy (Tokaji and Ogawa, 1990), and pure titanium (Tokaji, Ogawa, Kameyama, and Kato, 1990) revealed a similar overall growth behavior as crack growth rates were markedly decreased by grain boundaries, triple pints and interfaces between phases depending on the microstructures. These studies also supported the above mentioned research works that large decreases in microstructurally "short" crack growth rate are more frequent in fine grained materials than in coarse grained materials (see Fig. 21). This was observed to increase the overall fatigue life in the fine grained materials.

In fine and coarse grained materials such as in pure titanium (Tokaji, Ogawa, Kameyama, and Kato, 1990), the crack path was observed to be extremely tortuous and it increased with increase in grain size. Therefore, decrease in crack growth rate resulted. Fig. 22 shows that decreases in crack growth rate are observed more frequently in fine grained material as compared to coarse grained material. This was attributed to grain boundary and crack deflection. As described in one study (Suresh, 1983), the deflection in crack path might lower the crack driving force. Also, as hypothesized by some researchers (Tokaji and Ogawa, 1992), the tortuous nature of crack path morphology in pure titanium might be associated with planar slip characteristics or fewer slip systems than bcc and fcc

metals. Furthermore, they postulated that because of this, as cracks reached grain boundary, large changes of growth direction might occur due to the incompatibility of deformation and it was related to misorientation between two grains. Although extensive research studies on the microstructurally "short" crack behavior were performed on various possible materials, a direct comparison is not possible because of some obvious differences in test conditions viz. stress level, loading type and other test parameters. However, as indicated before, a general behavior of microstructurally "short" cracks is observed in many of the materials bearing a few contradictory results.

The effect of stress ratio (R) on microstructurally "short" cracks was studied by some workers (Tokaji and Ogawa, 1990), The microstructural effect at R=-1 and at R=0 is shown in Fig. 23. As can be seen from this illustration, in aluminum alloy (7075-T6) tested at R=-1 and R=0, faster growth was recorded at R=-1. Also, the results at R=-2 showed the fastest crack growth rates. (Tokaji and Ogawa, 1992) observed stage I facets on the fracture surfaces when tested at R=-1 and R=-2 but not at R=0. They related the growth behavior of microstructurally "short" cracks to the existence of stage I facets. Also, some studies on Stage I crack formation as related to "short" crack growth behavior are discussed in the next section.

2.1 Stage I crack formation studies and its relation to "short" crack growth behavior

The slip mechanism for the formation of fatigue cracks was first studied by Ewing and Humphrey, Gough and Forsyth among others. Since then, formation of persistent slip bands (PSB) has been recognized as a general phenomenon for the nucleation of fatigue cracks. Forsyth termed the formation and growth of crack in slip bands during the fatigue process as Stage I. Stage I or slip band cracking was related to the range of resolved shear stress on the slip plane (Forsyth, 1969). Also, slip bands produced by cyclic stress were shown to be a series of grooves and ridges and the fatigue deformation mechanism of "slip band intrusion and extrusion" was related to Stage I crack growth mechanism (Forsyth, 1969). However, not all the fatigue cracks form from slip bands. Under favorable conditions of stress and environment these cracks may form on those planes most closely aligned with the maximum shear-stress directions in the component or test specimens. Some "short" crack research studies related the behavior of cracks in the "short" regime to Stage I cracks and they are discussed below.

Two high strength and low alloy steel containing V and Nb were fatigue tested to observe the nucleation of cracks and growth of "microcracks" (Kim and Fine, 1982). Strain controlled fatigue tests were conducted in 30-40% humidity air at room temperature. Also, results of crack formation at different mean stress were reported. This study showed that at all stress levels, fatigue cracks were

observed to form along persistent slip bands. The number of cycles to nucleate a crack of size 5 μm long was observed at 2000 to 3000 magnification and the "microcracks" were always found to be associated with slip bands. (Kwun and Fournelle, 1982) also reported more density of slip bands and "small" cracks for quenched and tempered at a lower temperature when compared to tempering performed at higher temperature in Niobium bearing high strength and low alloy steel.

Although the formation of cracks in extruded aluminum alloy X7091 containing Zn-Mg-Cu and Co occurred at grain boundaries at both low and high stresses, the same was not true when the material was subjected to thermomechanical treatment. This resulted in the slip band crack nucleation (Hirose and Fine, 1983). Another study (Kim, Mura and Fine, 1978) in 4140 steel also showed that "microcracks" formed at grain boundaries in as-quenched specimens and at intrusions and extrusions in the tempered specimens.

In another study, Tokaji and Ogawa, 1988, observed many straight lines on the facets of medium carbon steel and the direction of these lines were confirmed to be consistent with the slip direction <111> of this material. This indicated that the crack grew along slip planes in a shear mode and it was related to stage I crack growth mechanism. Moreover, they (Tokaji and Ogawa, 1992) argued that the microstructural effect "short" crack growth occurred as a sequence of stage I crack growth.

At higher strains, at two different microstructures, when fatigue tested, Ti-6Al-2Sn-4Zr-6Mo exhibited nucleation of cracks within the slip bands in Widmanstatten plus grain boundary alpha and equiaxed structures of different alpha particle (Mahajan and Margolin, 1980). However, at low strains the crack nucleation occurred at alpha-beta interface and the resultant "microcracks" linked up and extended.

Similar observation was made in another study conducted in 2024 and 2124 aluminum alloy in the T-4 condition (Kung and Fine, 1979). This study was conducted in dehumidified laboratory air, 10% humidity as well as 50% humidity. The loading was applied in tension-tension and also in tension-compression with the direction of stress normal to the long direction of the notch and parallel to the specimens rolling direction. It was found that at high stresses the fatigue cracks formed on "coarse slip lines" in both alloys. However, at low stresses majority of cracks originated from the constituent particles. However, the important finding of this study was the probability of nucleation of a fatigue crack at a constituent particle size normal to the stress direction decreased below 6 μm and the crack formation mode was along slip bands which originated from the inclusions.

Polycrystalline copper of commercial purity (99.9%) exhibited formation of extrusions and intrusions along persistent slip bands within the grain and also in preferably oriented grain boundaries when tested in constant strain amplitude cyclic loading (Polak and

Liskutin, 1990). "Short" crack growth was characterized by PSB nucleation and was related to localization of the cyclic slip at or close to grain boundaries and "the cracks advanced by linking a newly nucleated crack at the tip of the existing crack or in front of it" (Plumtree and O'Connor, 1991) also observed the stage I crack growth to a depth of about 250-350 μm in Al-Mg-Si alloy subjected to two different processing techniques viz. extruded and squeeze-cast. They stated that the "short" crack behavior was observed during this portion of fatigue life.

Different kinds of crack nucleation mechanisms were observed in nickel base super alloy namely, Waspaloy (Yates, Zhang and Miller, 1993). Four point bending fatigue tests were conducted on Waspaloy specimens and was found that crack formed directly from PSBs, along a twin boundary and also in grains. Moreover, it was observed that in all the three types, cracks formed at 45° to the principal stress axis suggesting the Stage I growth until the crack depth of around 600 μm. This study proposed a model for the "short" crack growth behavior in Waspaloy incorporating the microstructural effect as well as the characteristic nature of grain boundary in blocking the growth of "short" crack.

3. Environment effects on "Short" crack behavior of materials

The environment (chemical) and temperature effects on the fatigue crack propagation in the "long" crack regime are not well studied and the same is applicable to "short" crack studies. Very few researchers have performed "short" crack studies to evaluate these issues.

In earlier study, very low environmental influence on "small" surface cracks was observed in a 7075-T651 alloy when tested in ambient air and in purified nitrogen (2 ppm water) (Lankford, 1983). However, some studies (Zeghloul and Petit, 1985, Petit and Zeghloul, 1986) have shown a greater effect of environment on the propagation of "short" through cracks grown in a 7075-T651 and T7351 when tested in air and in purified nitrogen and also the test results were compared in vacuum (see Fig. 24). Moreover, in another study (Petit and Zeghloul, 1990), faster growth rate of "small" surface cracks was observed in ambient air compared to vacuum when 7075-T651 and T7351 were fatigue tested. This behavior was related to water vapor embrittlement as in the case of long cracks. These studies have shown that the growth in ambient air of stage II cracks in a 7075 alloy can be rationalized with that of long and "short" through-section cracks in terms of ΔK_{eff} after correction for local plasticity. However, another study (Petit and Kosche, 1992), showed that the initial propagation in vacuum of "small" surface cracks naturally "initiated" on smooth specimens in 7075-T651 and

7075-T351 alloys is much faster than stage II propagation, and is similar to the intrinsic stage I regime, as identified on Al-Zn-Mg single crystals.

More dramatic results were shown in a review paper (Gangloff and Wei, 1986) in which it was concluded that "small" corrosion fatigue cracks in high strength steels when cycled in aqueous hydrogen producing environments grew up to 500 times faster than "long" crack at constant ΔK. From this paper, results illustrating the effect of crack size on corrosion fatigue in steels were reproduced and they are given in Fig. 25.

Another study (Akid and Murtaza, 1992) showed that environmental assisted "short" fatigue crack growth could influence the propagation of cracks beyond the "crack-arresting barriers" such as grain boundaries. Their studies in high strength spring steel using an intermittent fatigue tests in air and NaCl solution clearly showed that strain assisted dissolution caused the transition of stage I crack to stage II at shorter crack lengths. Also, (Boukerrou and Cottis, 1992), another study in a structural steel (BS4340 grade) in 3.5% NaCl and in pitting solution, showed that cracks could nucleate from corrosion pits when cycled at low stresses.

3.1 Temperature effects on "short" crack behavior of materials

At room temperature, "short" cracks were found to form primarily at inclusion particles, and less frequently at grain boundaries in Nickel base super alloy when cyclically loaded in four point bending specimens (Mei, Krenn, and Morris, 1993). However, when tested at elevated temperature (873°K) "short" cracks were observed to form from micropores, slip planes and carbide precipitates at grain boundaries in Nickel base super alloy (Okazaki, Tabata and Nohmi, 1990). In this study, Stage I fatigue fracture occurred on the {111} planes. Also, (Stephens, Grobowski and Hoeppner, 1993) slip band cracking was observed in Waspaloy tested at 25° C, 500° C and crack was observed to form at twin boundaries and slip bands at 700°C. Moreover, this study demonstrated the "faster" growth rate of "short" cracks at 500°C when compared to 25°C and 700°C. This was related to diffusive nature of slip bands and "changes in material leading to precipitate coarsening" at 700°C. Similar observation of slip band cracking was found in Waspaloy when tested at 19°C and 500°C (Healy, Grabowski and Beevers, 1991). In addition, "short" cracks formed at coarse carbide particles. The "short" crack growth rate measured at R=0.1 was "faster" at 500°C when compared to 19°C at an equivalent values of stress intensity range as reported by Stephens, Grobowski and Hoeppner. In another study (Suh, Lee and Kang, 1990), numerous "microcracks" were observed to form at grain boundaries in 304 stainless steel specimens tested at 538°C.

In general, the mechanisms for the "short" crack behavior at room and elevated temperature as postulated in the above mentioned studies are related to

• absence of closure effects,

• heterogeneous microstructure (responsible for statistical scatter in the "short" crack growth rate (Goto, 1993; Goto, 1994)),

• grain boundary cracking (may result from embrittlement due to stress-assisted grain boundary oxidation during heat treatment),

• grain orientation (if the grains are favorably oriented the cracks are observed to grow "faster" and if not "short" crack growth either gets slowed down or arrested),

• crack deflection (may occur within a grain or when the crack passed to another grain),

• crack tip deflection resulting in "roughness induced crack closure" and subsequent reduction in crack growth rate,

• Stage I crack growth (Stage I crack growth increases as the crack length increases. However, as the crack tip neared a grain boundary the "short" crack growth rate decreased because of crack deflection due to secondary slip and finally fracture occurring on the {111} planes in fcc).

4. Conclusion

This literature review on microstructural and environmental effects of "short" crack behavior of structural materials has clearly revealed the lack of experimental studies to characterize materials response in the "short" crack regime under fretting fatigue and corrosion conditions. As mentioned in this report, very little data have been published with regard to "short" crack formation mechanisms in corrosive environments of aluminum alloys in aircraft structures. In addition to a few previous studies (Hoeppner, 1979, Reeves and Hoeppner, 1978, Saliver and Hoeppner, 1979 and Hoeppner and Krupp, 1974) in which pitting was modeled statistically with different materials and specimen types, recently, as discussed before in this report, there was a study demonstrating corrosion fatigue induced "short" crack formation from pits (Akid and Murtaza, 1992). Also, a recent study (Ma and Hoeppner, 1994) has shown that pits form in different shapes in contradictory to general assumption that pits have hemisperical shape. Although this assumption simplifies the modeling part of research (Kondo, 1989), further studies to characterize the formation of cracks from pits in the "short" crack regime must be evaluated as outlined in the proposal submitted by QIDEC (Hoeppner, 1995). Apart from these studies the literature search has not found any "short" crack studies to evaluate the formation of cracks from pits and their crack morphologies and paths. Moreover, this may further be aggravated by fretting mechanism(s) in conjunction with fatigue and corrosion. It is envisioned that part II of the report may provide additional

insight of the "short" crack behavior under synergistic conditions of fretting fatigue and corrosion in aluminum alloys. This research will be further expanded as mentioned in the proposal that more studies will be conducted to develop some basic understanding on pitting corrosion fatigue as well as corrosion fretting fatigue in aluminum and titanium alloys.

To conclude the report, a global view of the "short" crack challenge is given in appendix III.

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Appendix I

"Short" crack test results extracted from literature

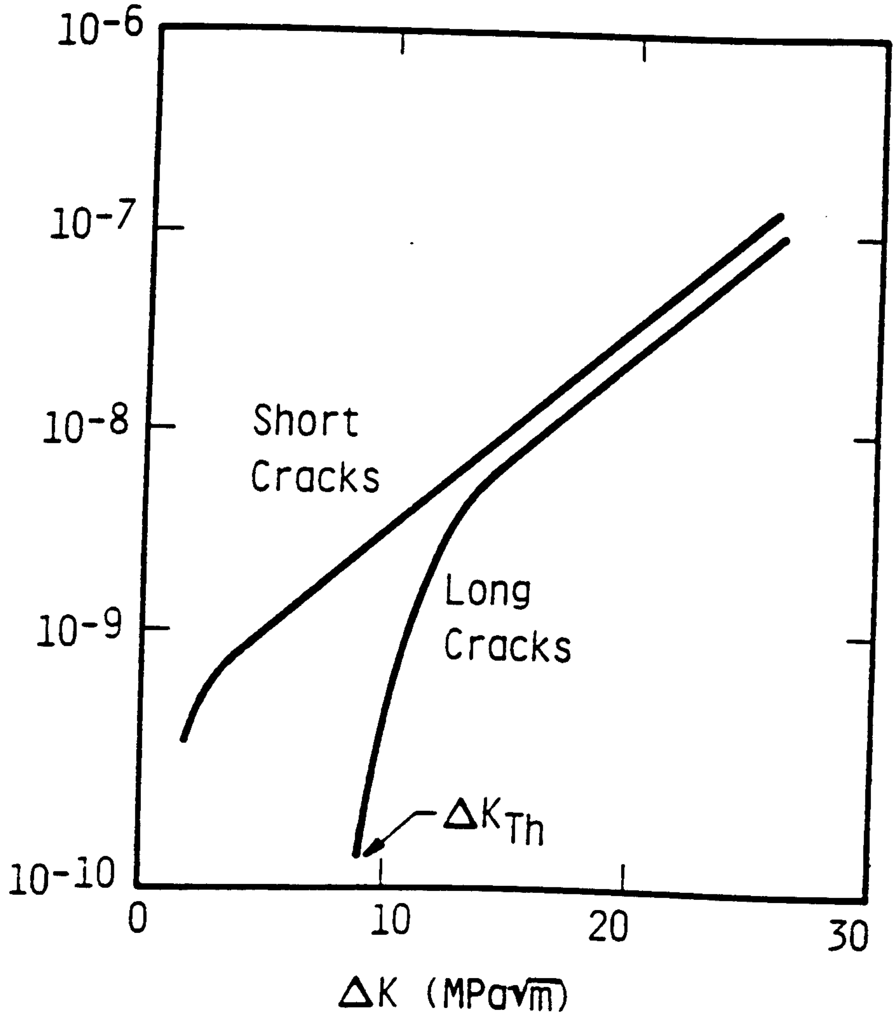


Figure 1. Fatigue crack growth rate versus cyclic stress intensity for long and short cracks in A508 steel (Breat, Mudry, and Pineau, 1983)

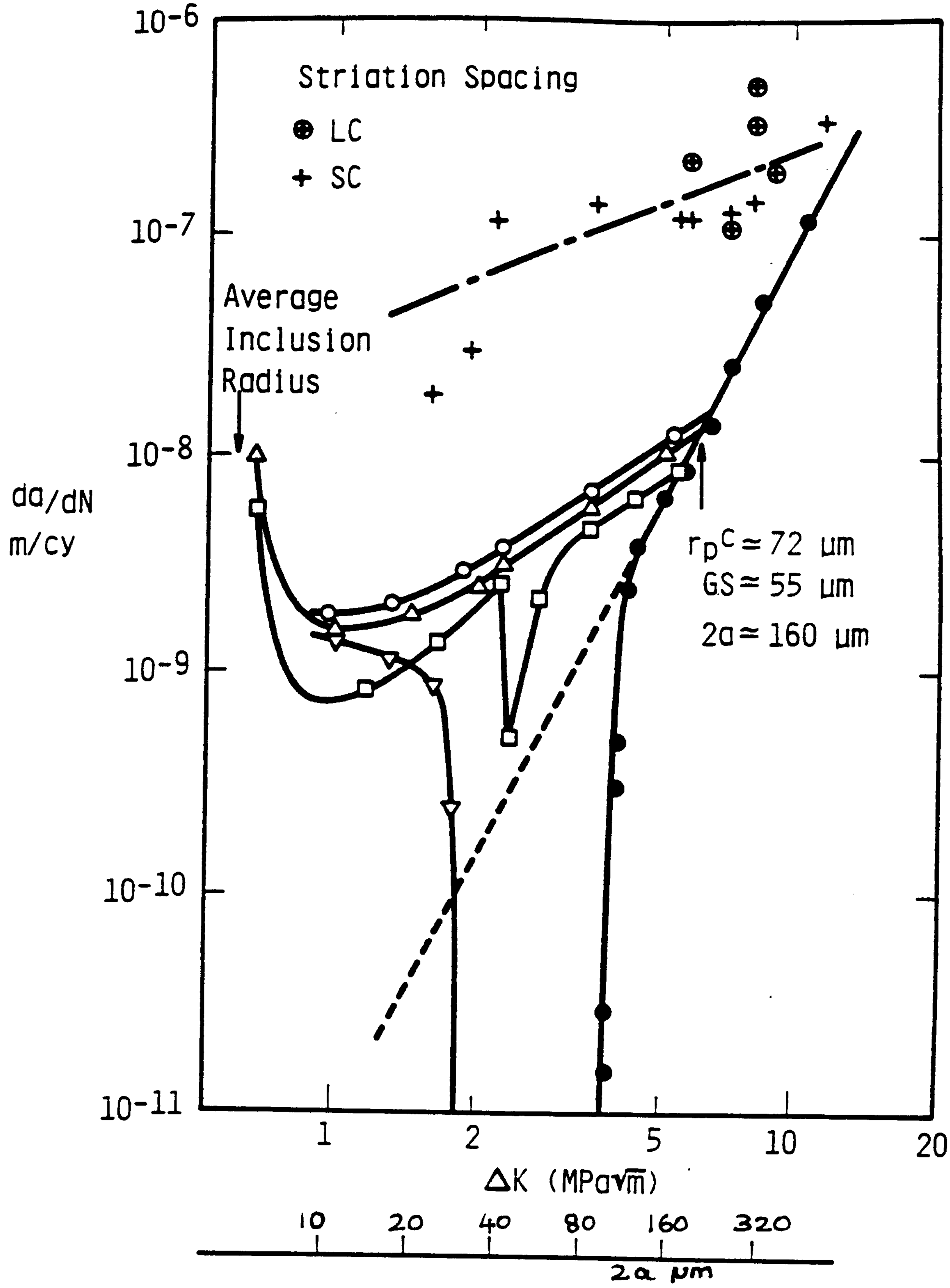


Figure 2. Crack growth rates for small and large fatigue cracks in 7075-T651 Al, versus ΔK and, for the small cracks, 2a (Lankford and Davidson, 1986)

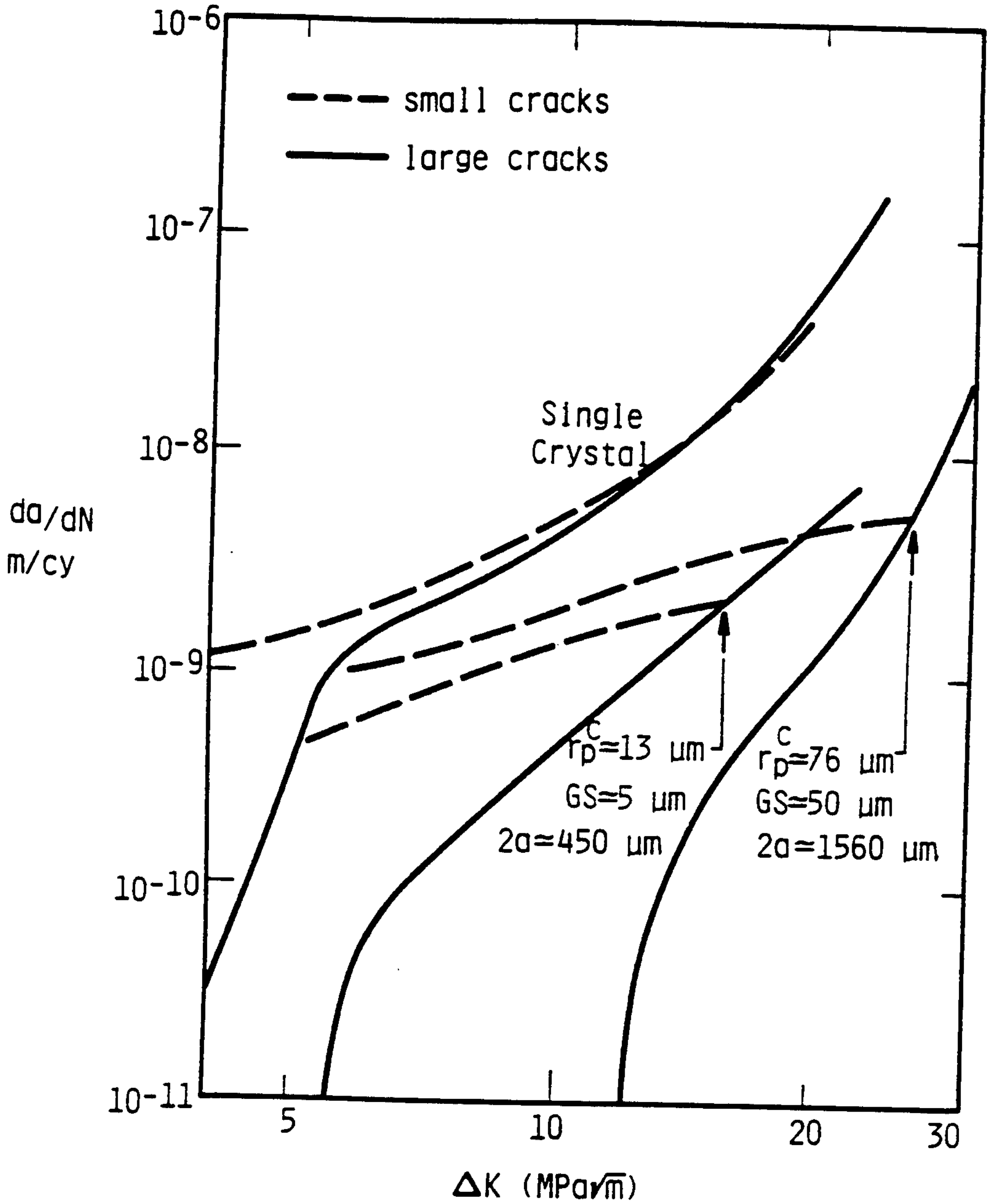


Figure 3. Crack growth rates for small and large fatigue cracks in fine grained (5 μm), coarse grained (50 μm), and single crystal Astroloy (Hicks and Brown, 1984)

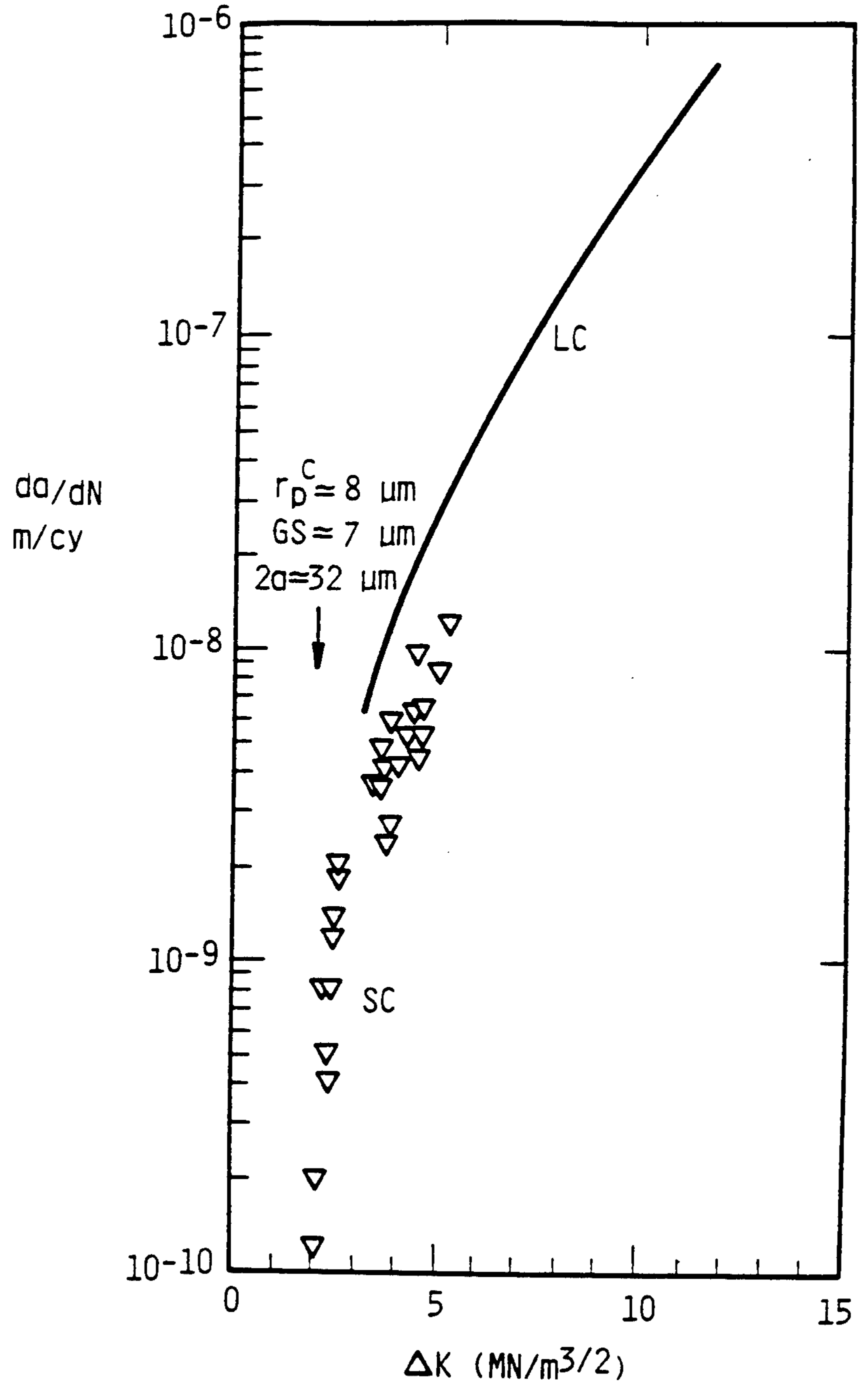


Figure 4. Growth of large and small cracks in fine-grained 7091 P/M alloy (Lankford and Davidson, 1986)

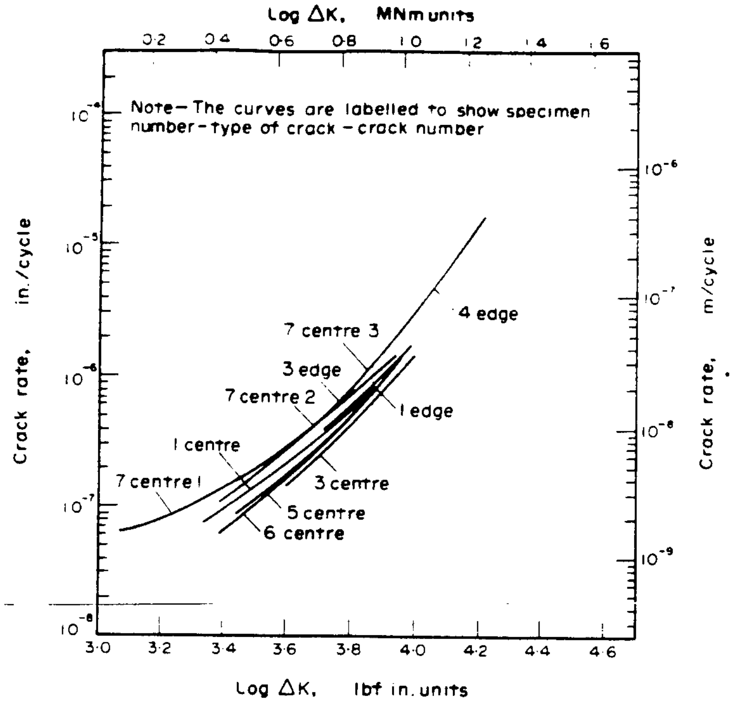


Figure 5(a) Propagation of very short cracks-Depth 0.006-0.5 mm (0.00025-0.02 in.) for Aluminum alloy BS L65.(Pearson, 1975)

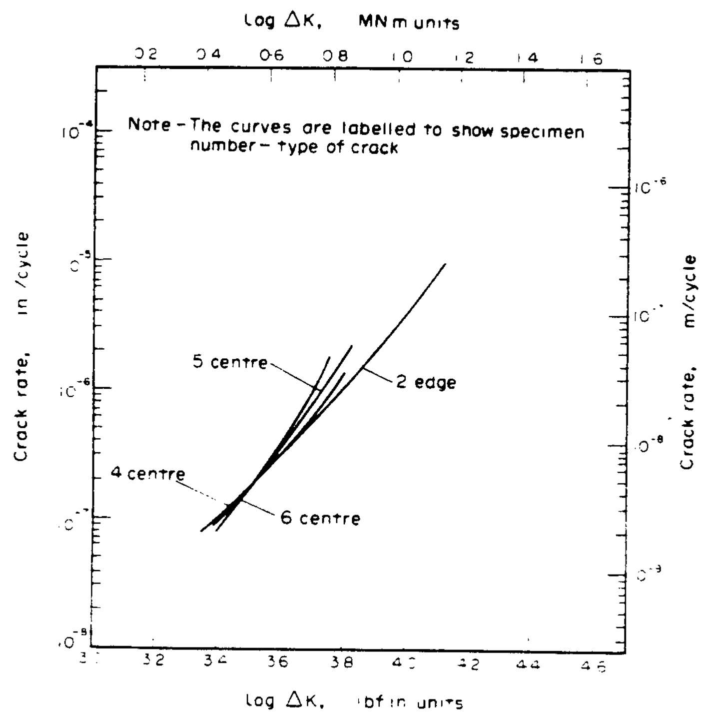


Figure 5(b) Propagation of very short cracks-Depth 0.012-0.5 mm (0.0005-0.02 in.) for Aluminum alloy DTD 5050.(Pearson, 1975)

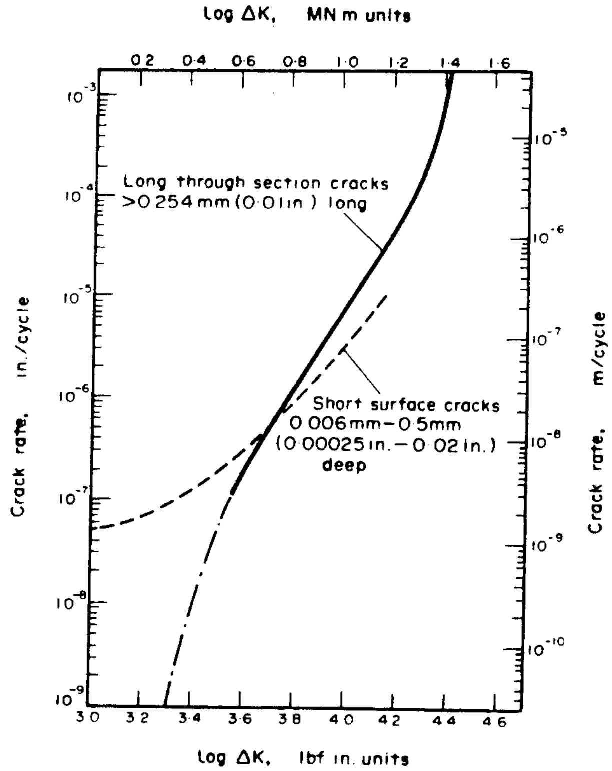


Figure 5(c) Fatigue crack propagation curves for long and short cracks. Aluminum alloy BS L65.(Pearson, 1975)

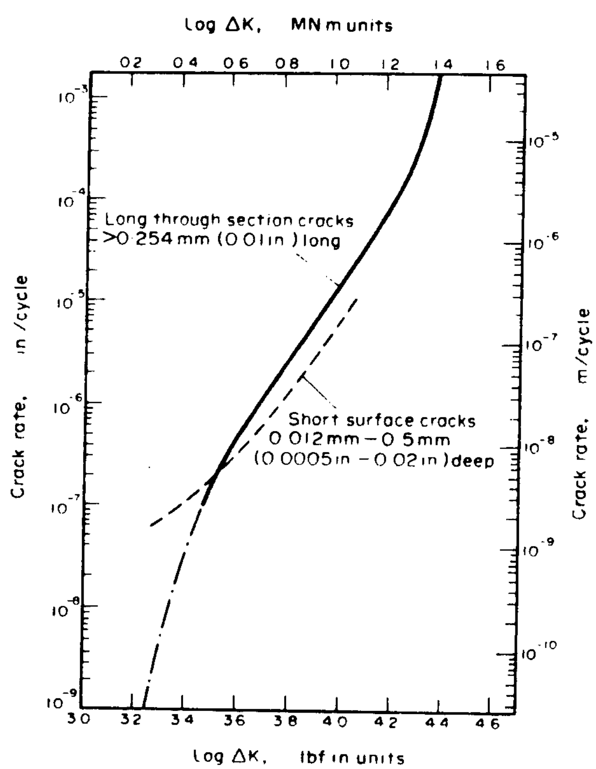


Figure 5(d) Fatigue crack propagation curves for long and short cracks. Aluminum alloy DTD 5050.(Pearson, 1975)

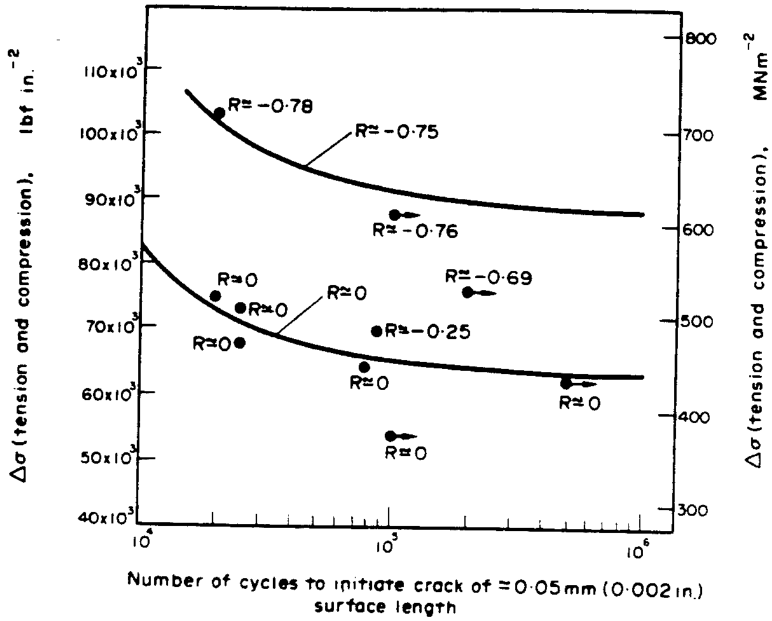


Figure 6. Number of cycles to initiate a crack of 0.05 mm (0.002 in.) Surface length plotted against ΔSigma for alloy BS L65 (Pearson, 1975)

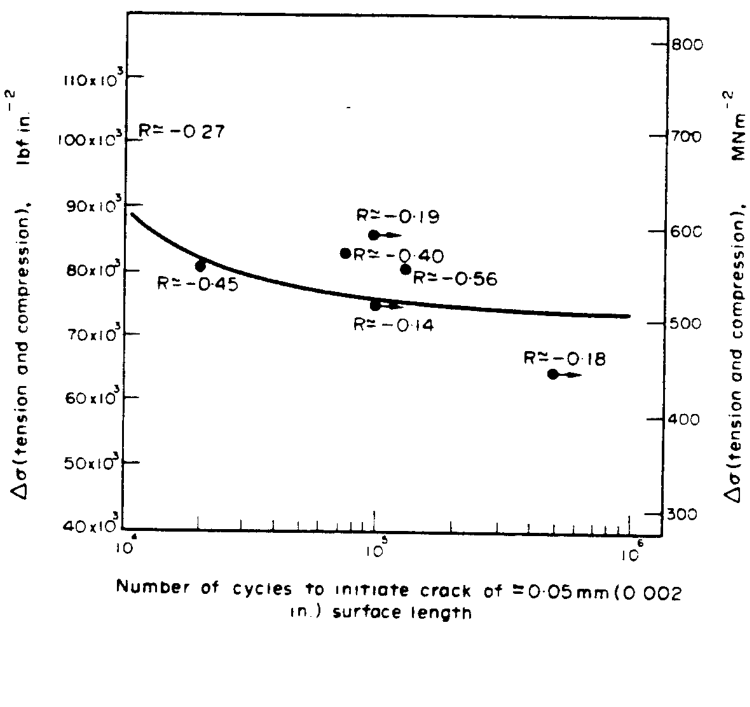


Figure 7. Number of cycles to initiate a crack of 0.05 mm (0.002 in.) Surface length plotted against ΔSigma for alloy DTD 5050 (Pearson, 1975).

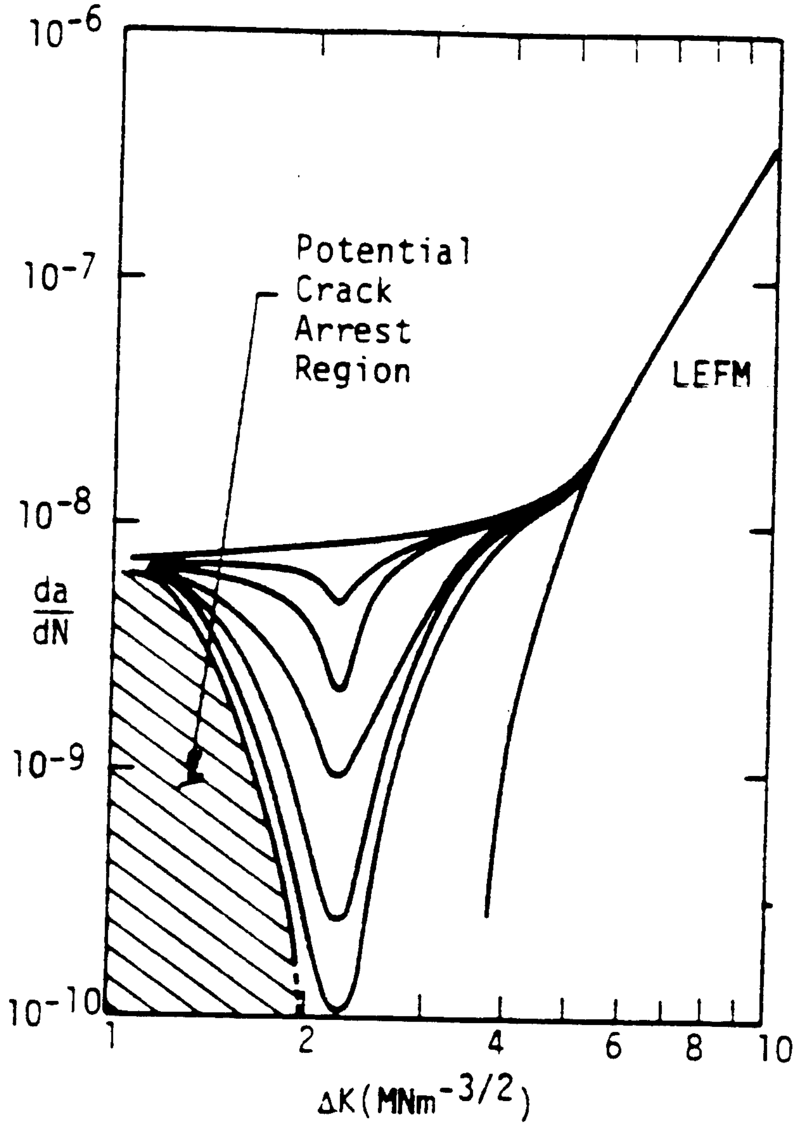


Figure 8. Schematic idealization of the results of "short" and long cracks. (Lankford, 1982)

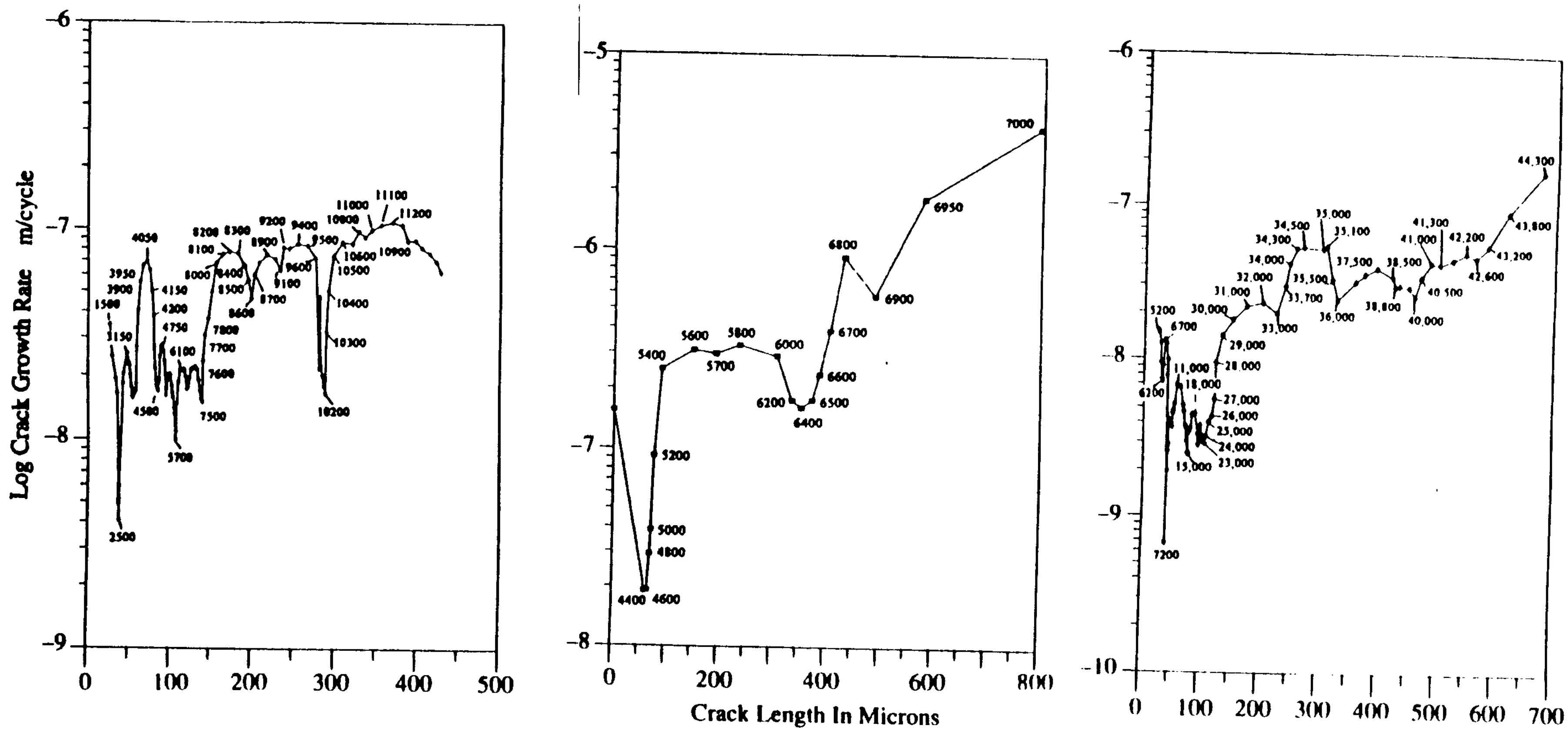


Figure 9. "Short" crack growth from corner notches: (a) Specimen 1; (b) Specimen 2; (c) Specimen 3. (Pan, De Los Rios and Miller, 1993)

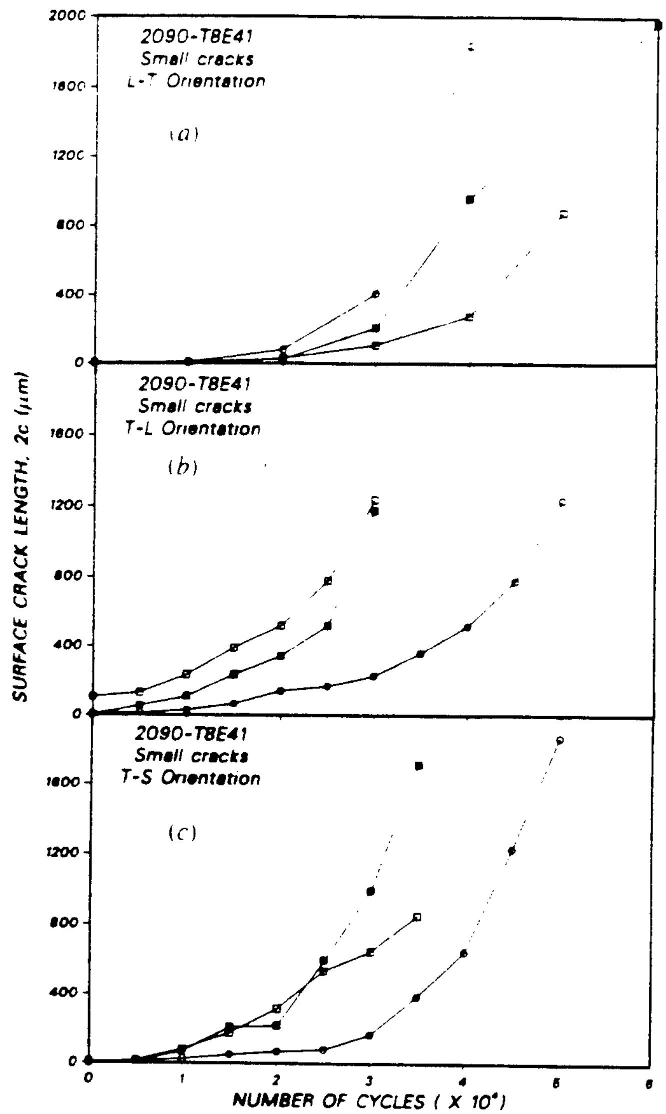


Figure 10(a) Replication data showing the surface crack length (2c) of "small" cracks in 2090-T8E41 alloy as a function of number of cycles, for (a) L-T. (b) T-L, and (c) T-S orientations. (Venkateswara Rao, Yu and Ritchie, 1988)

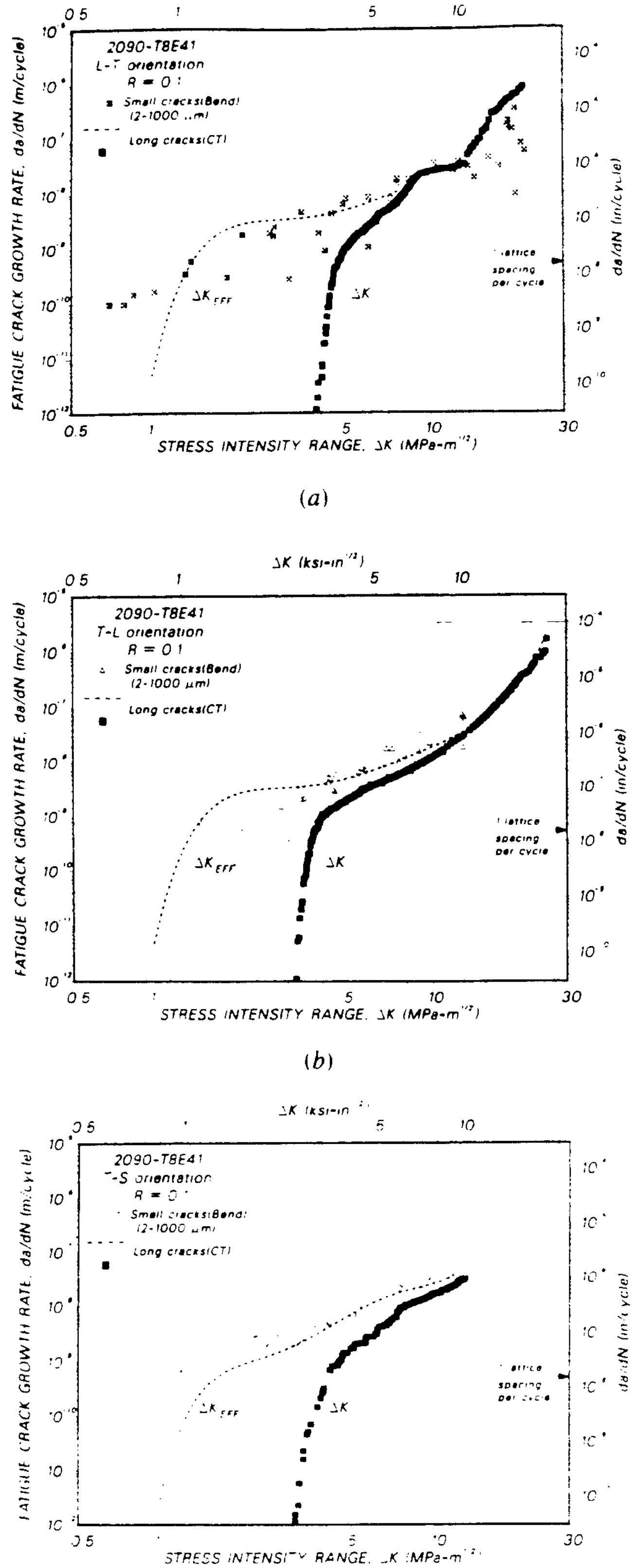


Figure 10(b) Comparison of the growth rates of long (> 5 mm) and naturally occurring small (2 to 1000 μm) fatigue cracks in 2090-T8E41 alloy, as a function of the nominal and effective stress intensity ranges, ΔK and ΔK_{eff} respectively. Data are presented for (a) L-T. (b) T-L, and (c) T-S orientations at R = 0.1. Note how "small" crack growth rates exceed those of long cracks by several orders of magnitude when compared on the basis of ΔK, yet show close correspondence when characterized in terms of ΔK_{eff}.
(Venkateswara Rao, Yu and Ritchie, 1988)

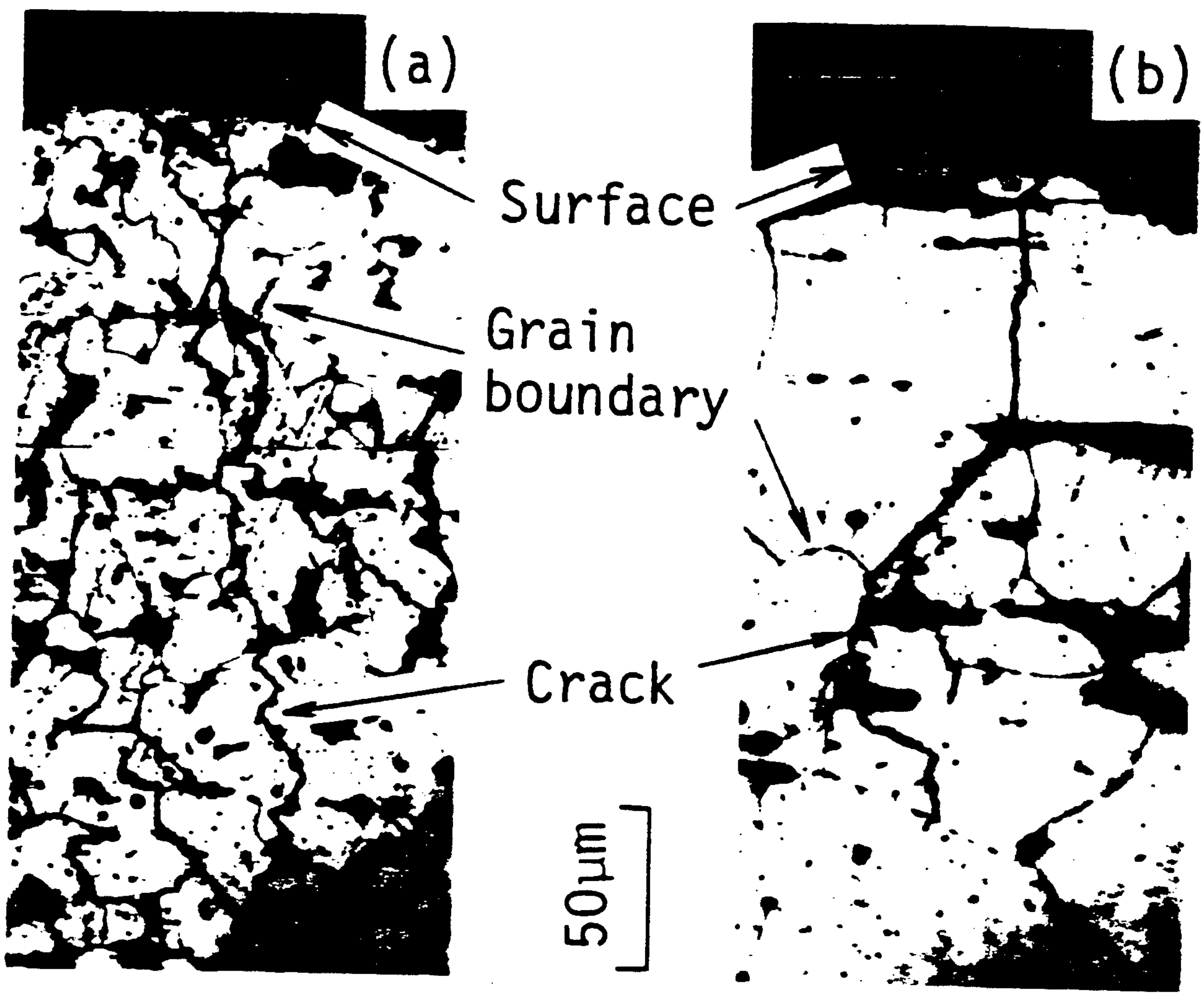


Figure 11. Macroscopic observations of "small" fatigue crack growth into the bulk. (a) Fine grained material (a = 0.276 mm). (b) Coarse grained material (a= 0.220 mm). (Tokaji, Ogawa and Harda, 1986)

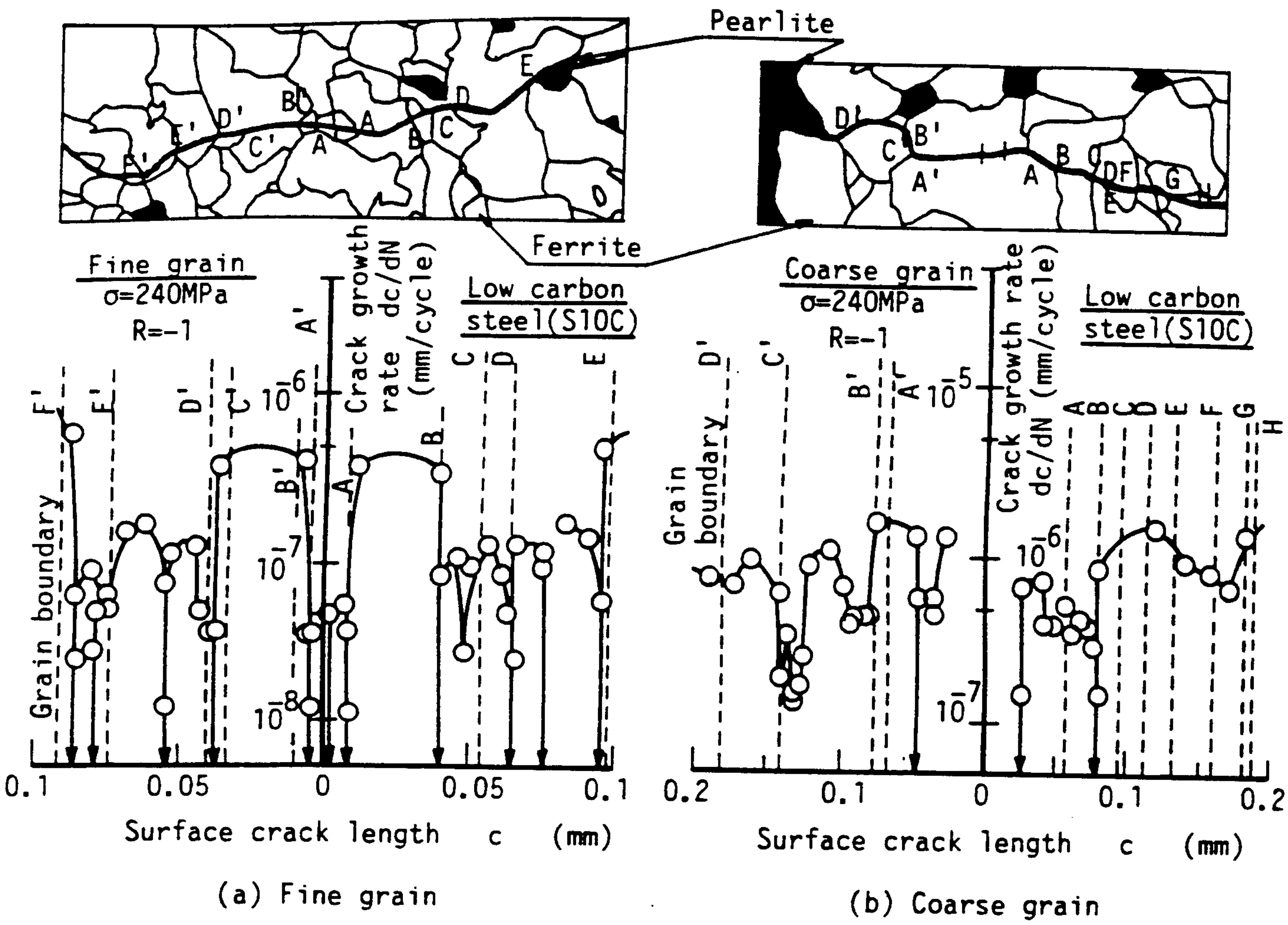


Figure 12. Effect of grain boundary on crack growth rate. (a) Fine grained material. (b) Coarse grained material. (Tokaji, Ogawa and Harda, 1986)

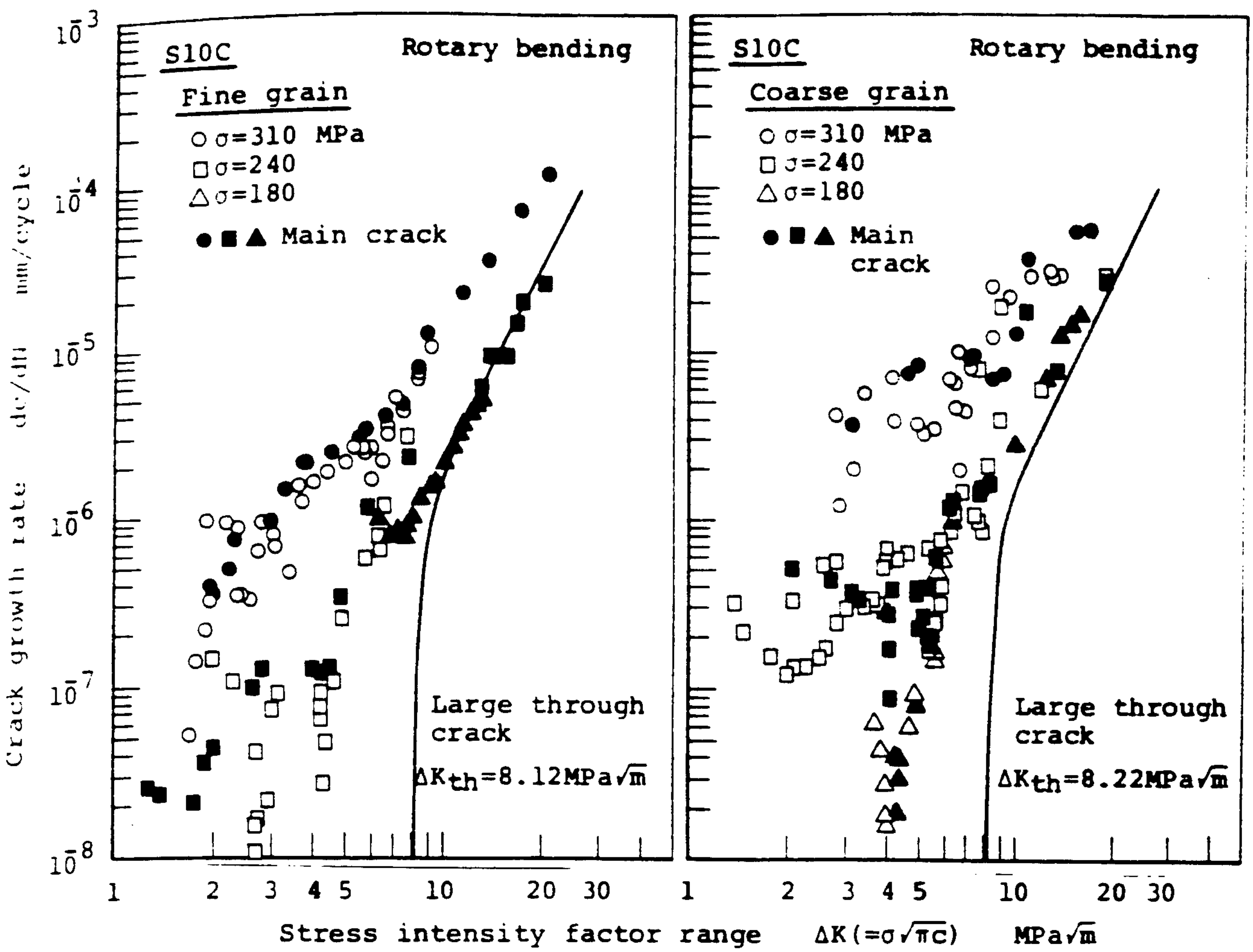


Figure 13. Relationship between growth rate and ΔK for "small" fatigue cracks. (Tokaji, Ogawa and Harda, 1986)

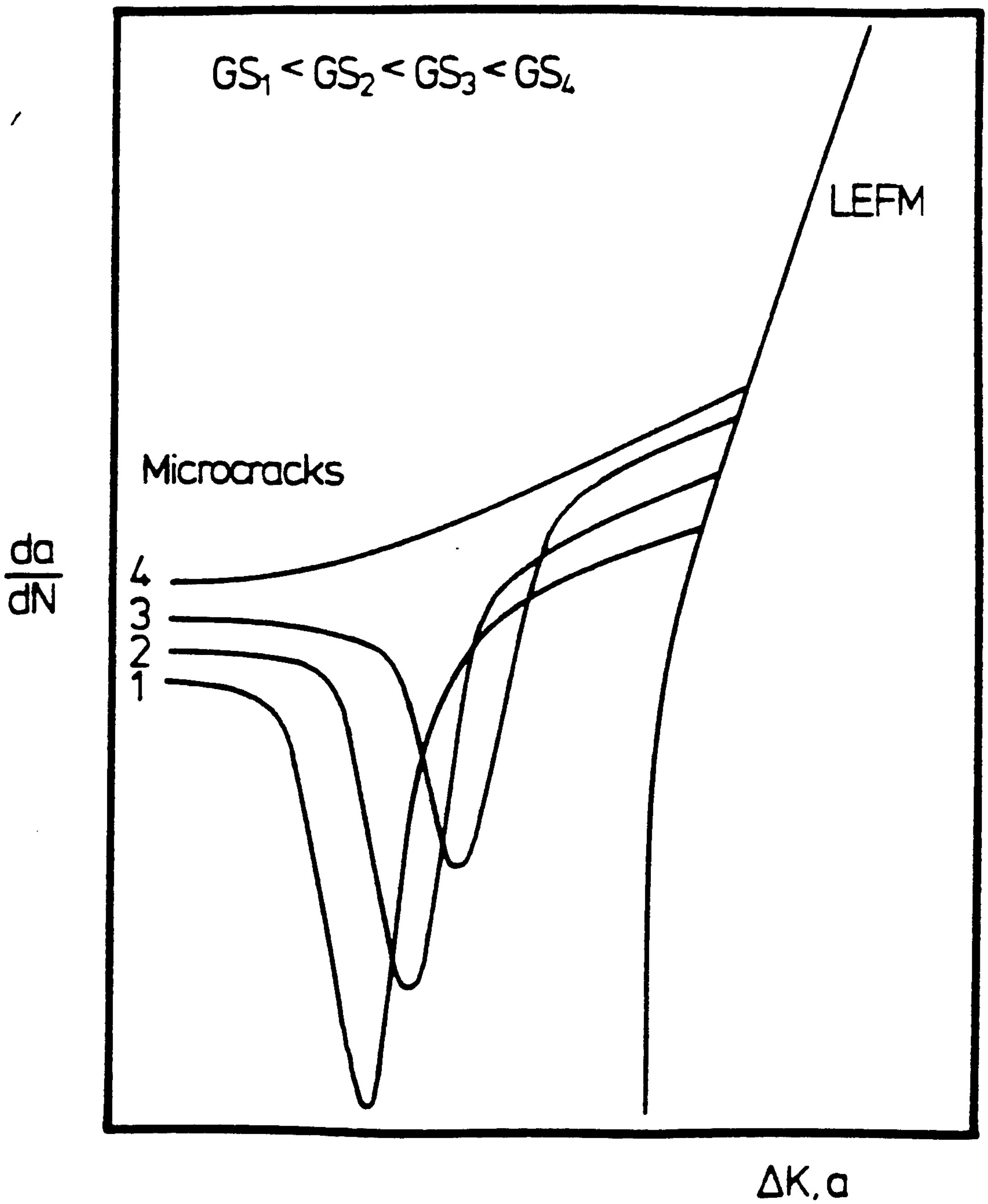


Figure 14. Schematic representation of the predicted effect of grain size on short crack growth (Lankford, 1982)

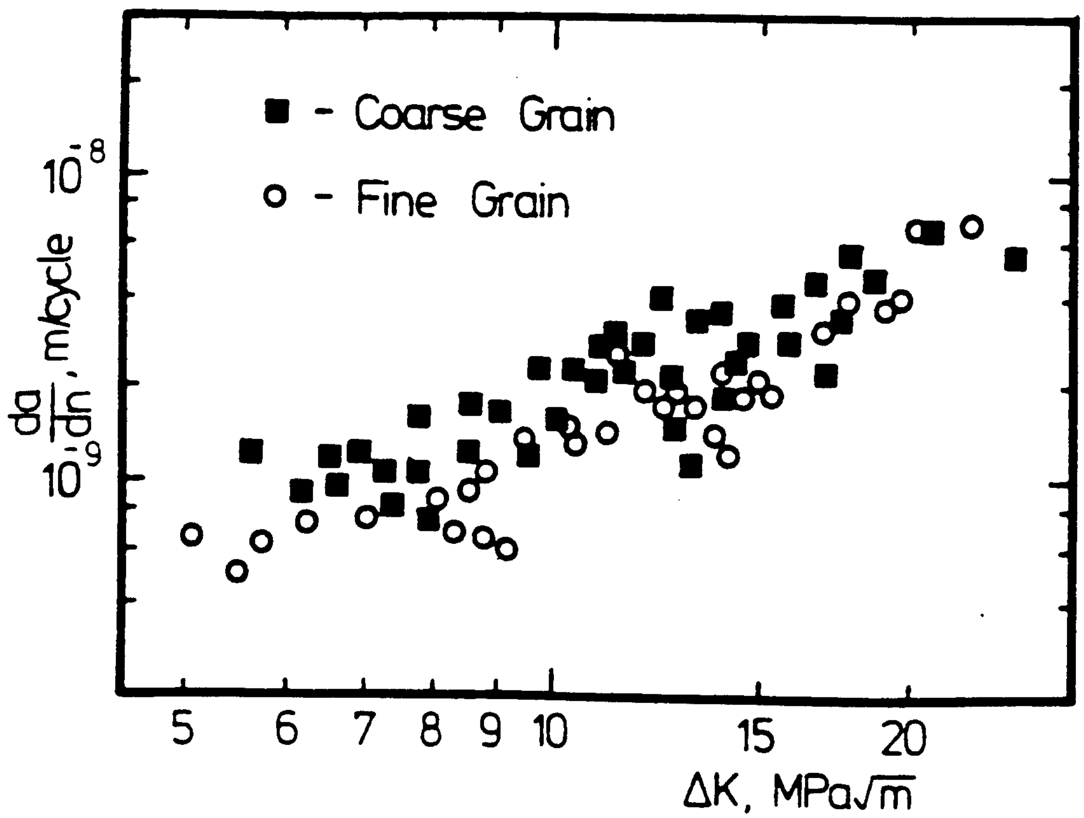


Figure 15. The effect of grain size on the fatigue crack growth rates in Astroloy (Brown, King, and Hicks, 1984)

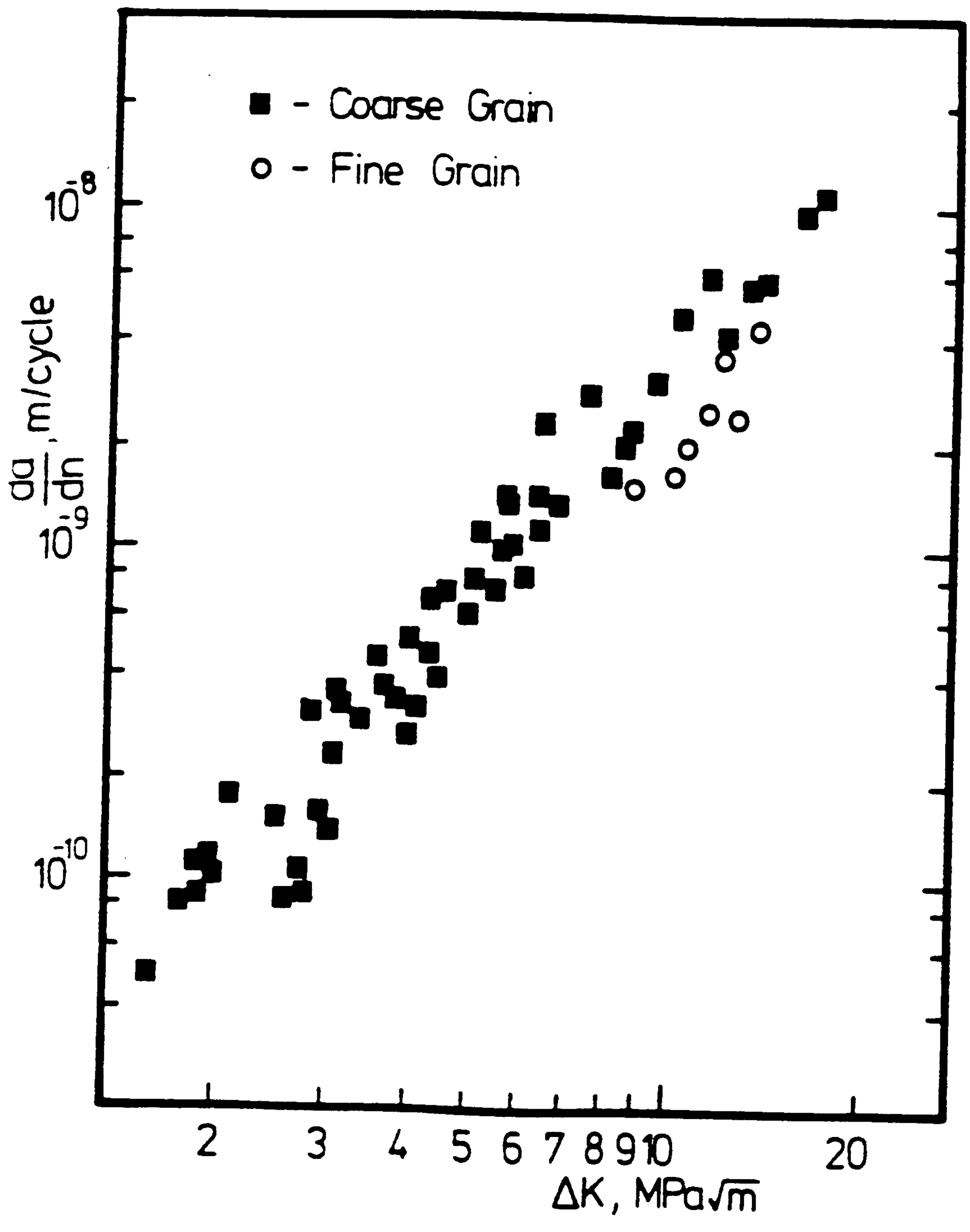


Figure 16. The effect of grain size on short fatigue crack growth in 0.2 wt% C steel. Coarse grain - 20.5 and 55μm, Fine grain - 7.8μm (Taira, Tanaka, and Hoshina, 1979)

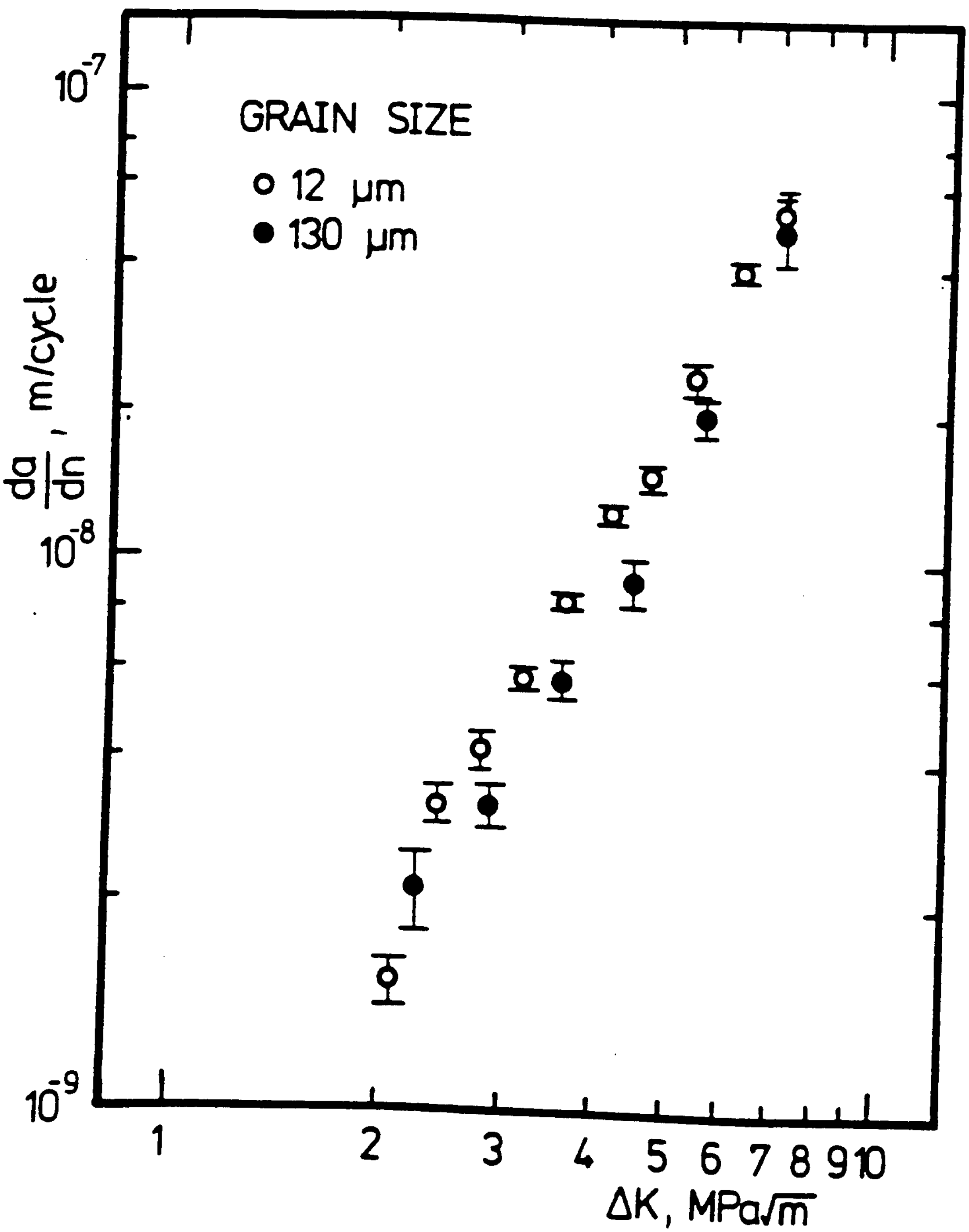


Figure 17. The effect of grain size on short fatigue crack growth in aluminum alloy 7075-T6 (Zurek, James, and Morris, 1982)

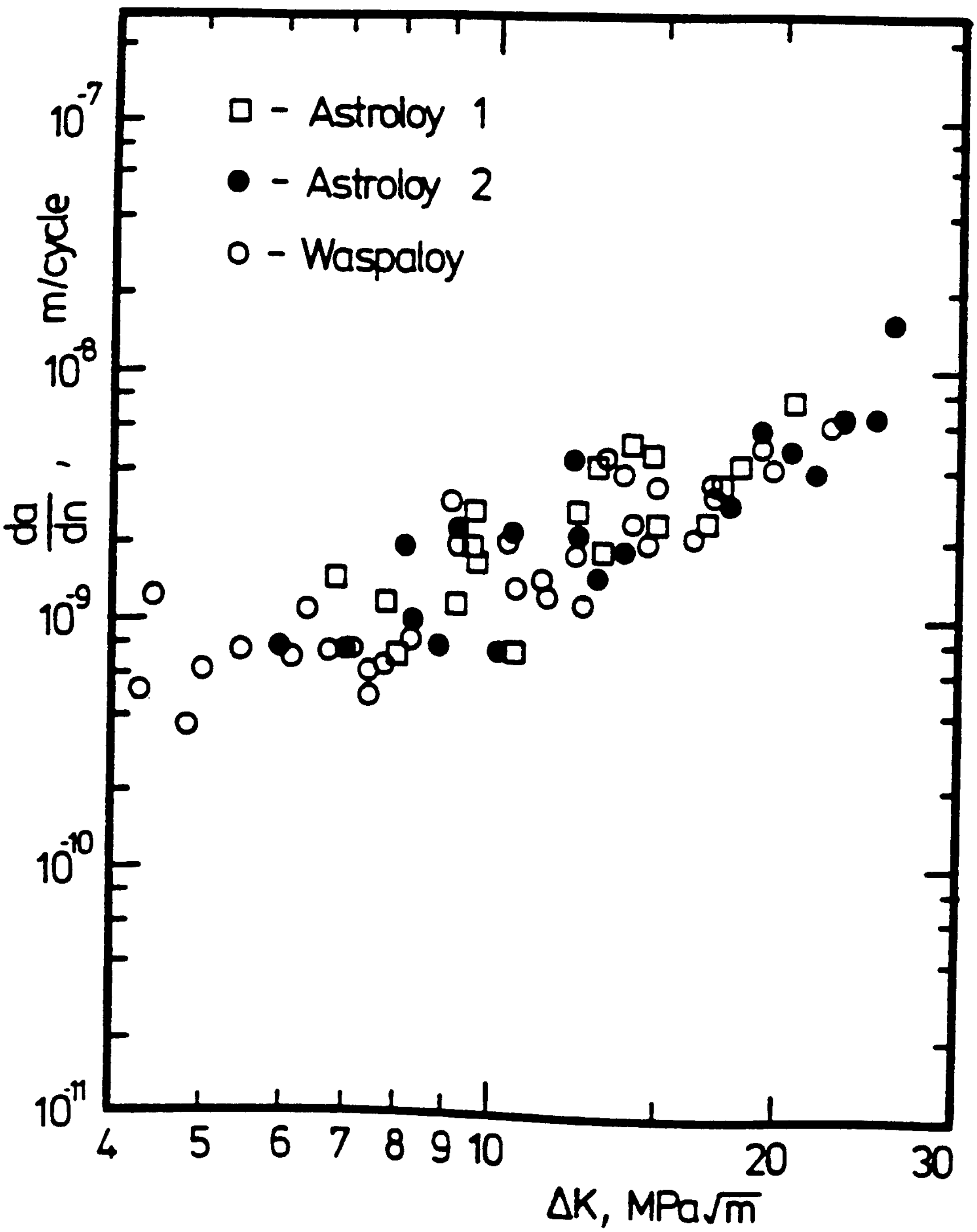


Figure 18. The effect of precipitate size and volume fraction on short crack growth in nickel-base superalloys (Brown, King, and Hicks, 1984)

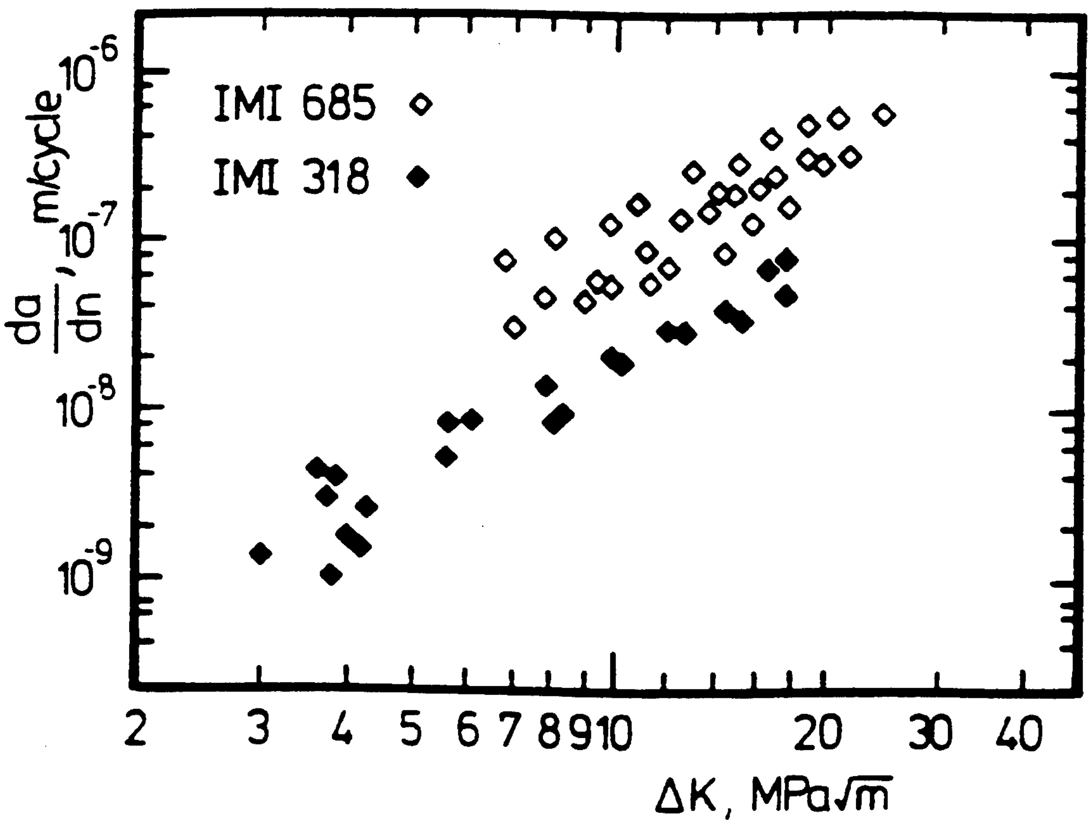


Figure 19. The effect of phase distribution on short crack growth in titanium alloys (Brown and Hicks, 1984)

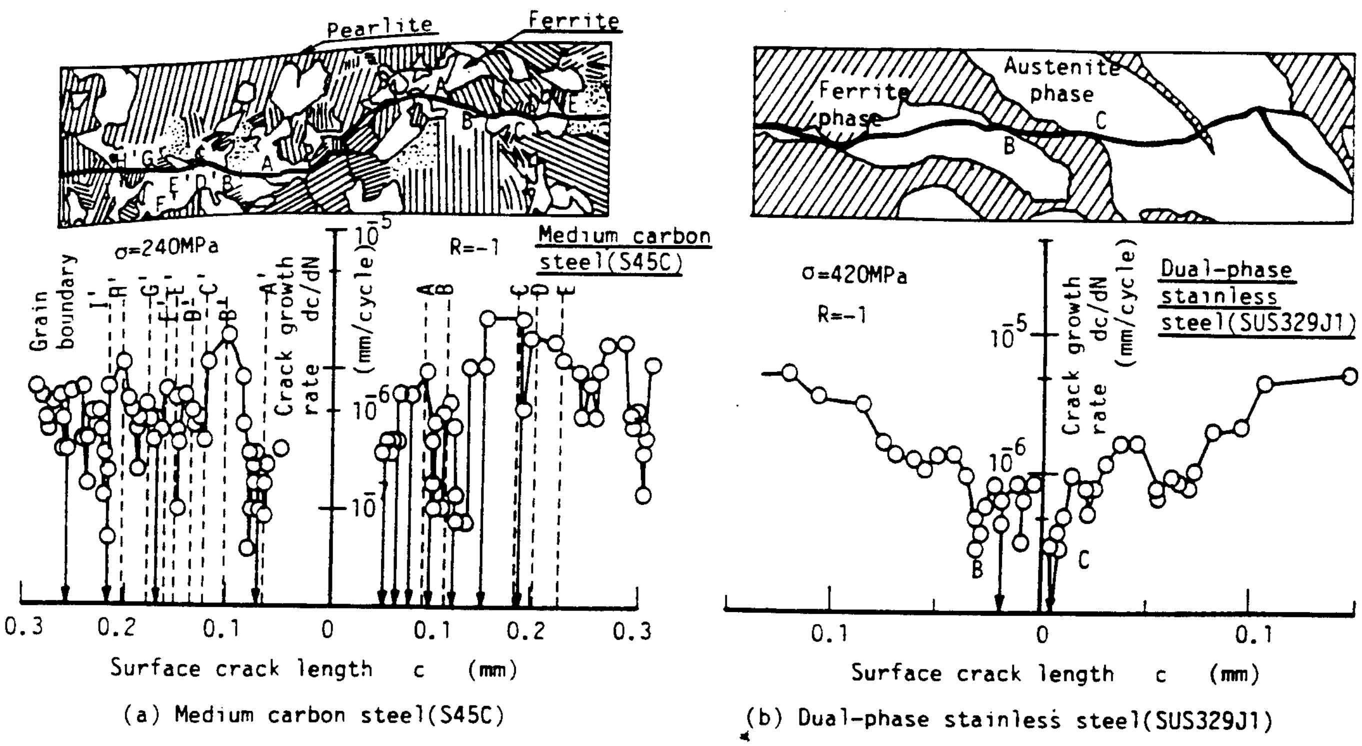


Figure 20. Growth behavior of microstructurally small cracks in metals with composite microstructures (Tokaji and Ogawa, 1988).

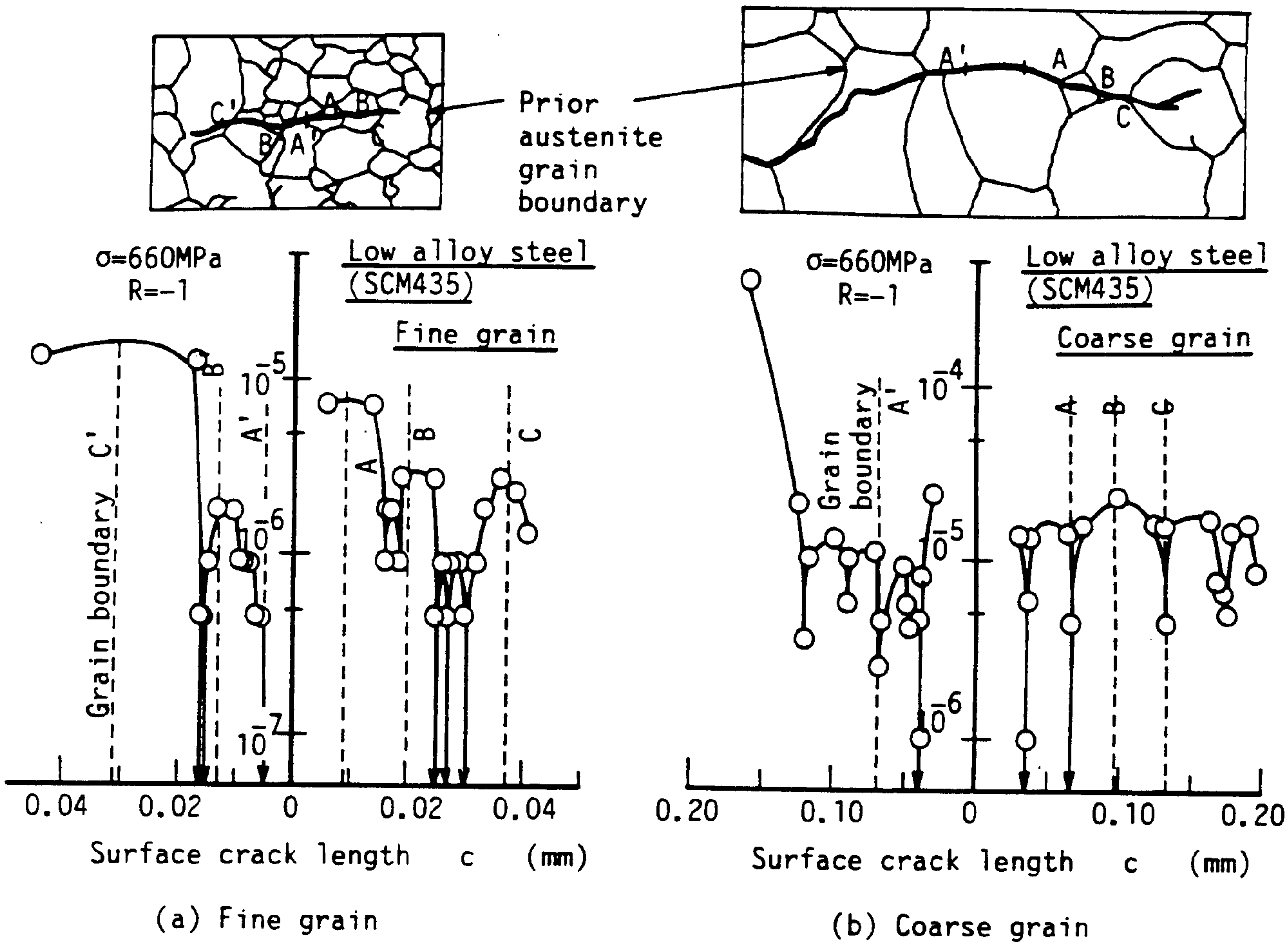
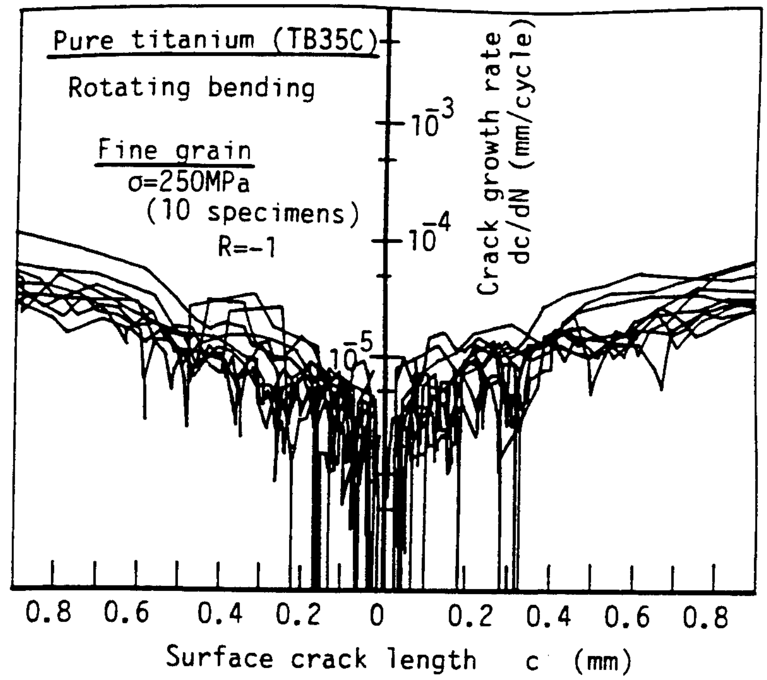
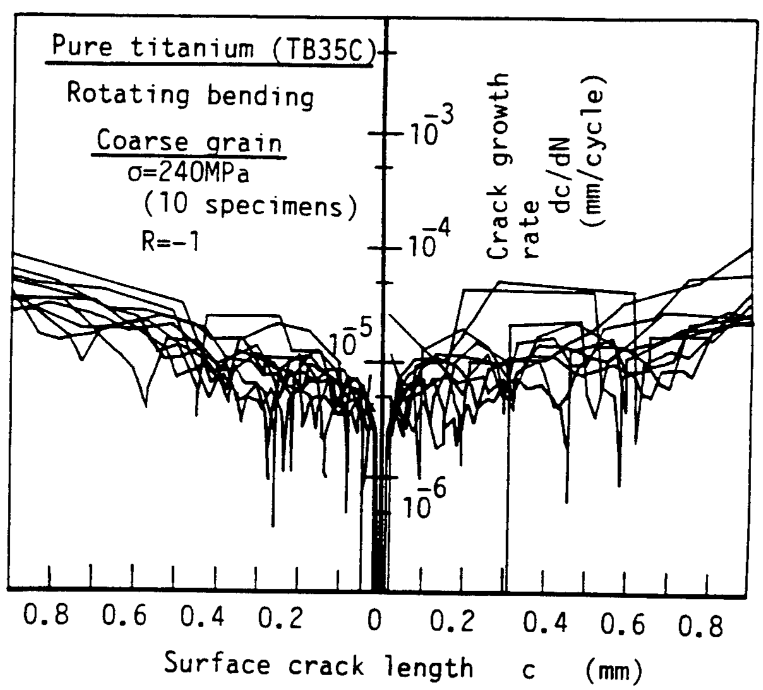


Figure 21. Growth behavior of microstructurally small cracks for fine and coarse grained materials in low alloy steel (Tokaji, Ogawa, Harada, and Ando, 1986)



(a) Fine grain



(b) Coarse grain

Figure 22. Growth behavior of microstructurally small cracks for fine and coarse grained materials in pure titanium (Tokaji, Ogawa, Kameyama, and Kato, 1990).

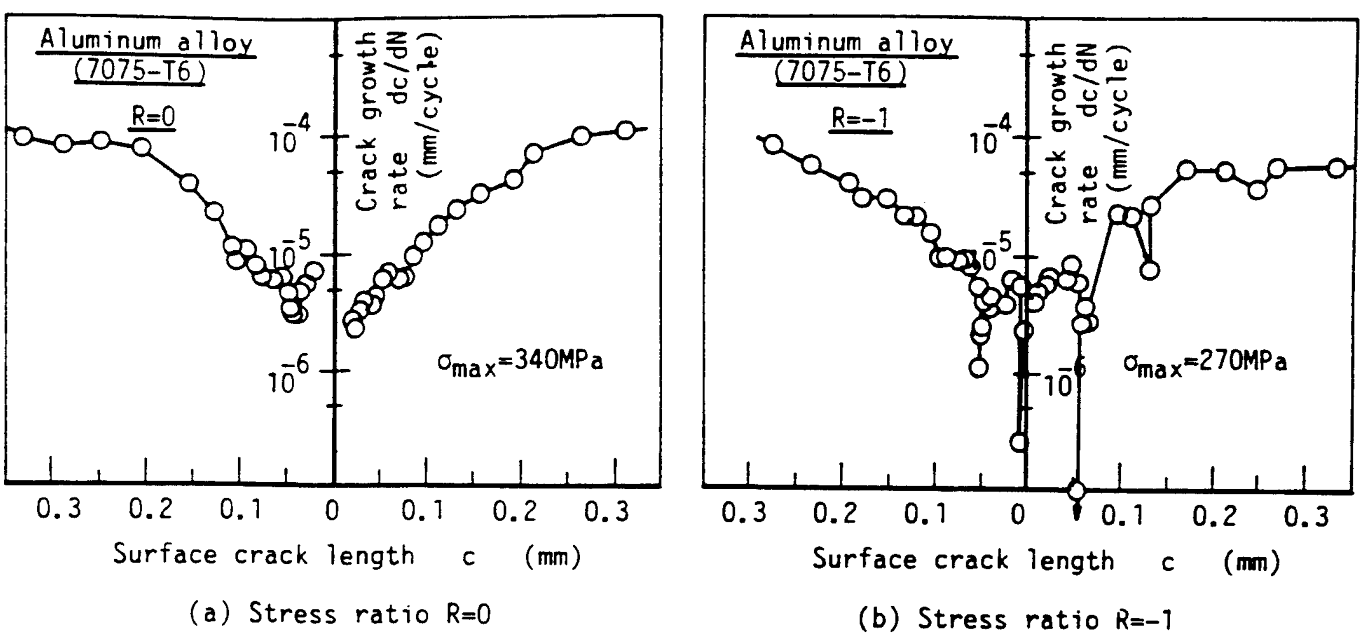


Figure 23. Growth behavior of microstructurally small cracks at two stress ratios in aluminum alloy 7075-T6 (Tokaji and Ogawa, 1990)

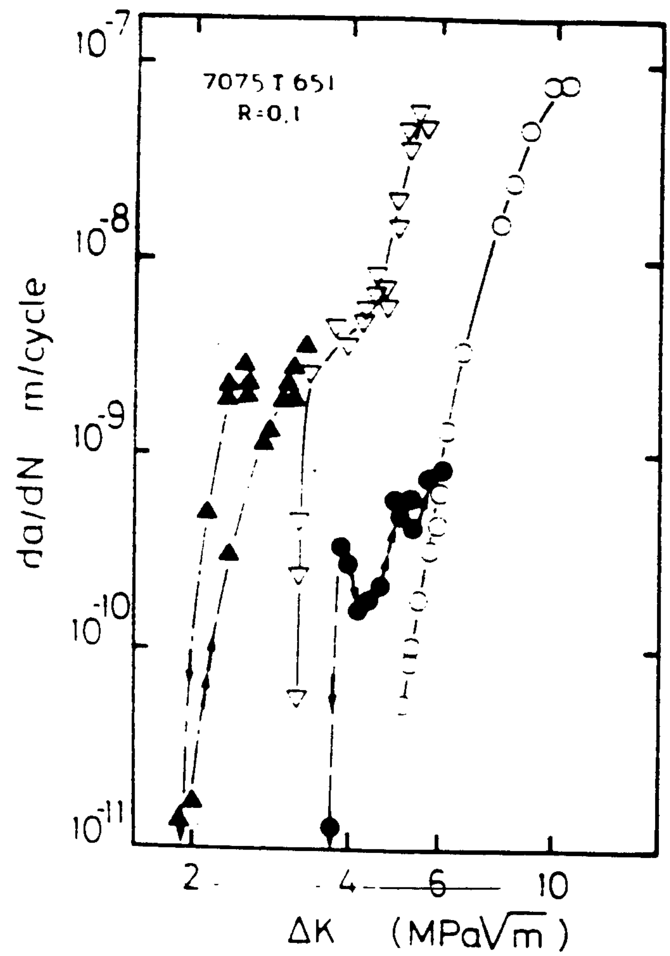


Figure 24. —*Short through crack data in air* (▲) *and vacuum* (●) *compared to long crack data in air* (∇) *and vacuum* (○) *for 7075 T651 CT specimens tested at* R = *0.1 and 35 Hz* (Petit and Zeghloul, 1986)

Material	σ_{YS} (MPa)	Environment	Crack Size Type (mm)	da/dN Defined by ΔK?	$\frac{da/dN_{SMALL}}{da/dN_{LONG}}$	Limiting Small Crack Size (mm)	Comments	Reference	
AISI 4130 (.3C-.9Cr-.2Ni)	1330	3% NaCl	.1-40 Long 1.9 Wide Edge, Elliptical	Yes - Air, Vacuum No - NaCl	1.2 to 500	>2.5	Growth Retarded by Increased K_{max}, R; $1/\nu_{max}$ Describes da/dN. 15 μm Grain Size.	13, 21	High Strength
ASTM A289-B (18.2Mn-4.8Cr-.6C)	1120	655 kPa H_2 Sat. H_2O 80°C	.1-40 Long 25 Wide Edge @ Notch	Yes - Air 23°C Yes - 10 Hz H_2 No - .02 Hz H_2	1.0 to 100	.8	Small Crack Growth Enhanced by Increased Initial ΔK and Loading Time. 100 μm Grain Size.	26	High Strength
HY130 (.1C-.5Cr-5Ni)	930	3% NaCl	.1-40 Long 1.9 Wide Edge	Yes - Air No - NaCl	4	>1.0	Initial ΔK Not Important.	22	Medium Strength
	972	3% NaCl	.4-40 Long 7.6 Wide Edge	Yes - Air No - NaCl	2	1.1	Limiting Crack Size Unaffected by $\Delta\sigma$, R.	18, 27	Medium Strength
	972	3% NaCl	.4-40 Long 7.6 Wide Edge	Yes - Air No - NaCl	1.0-1.8 Free Corrosion 0.8-1.3 Cathodic	0.9-1.4 5	Limiting Crack Sizes Up to 10 mm at Constant ΔK. No Significant Crack Closure Measured.	28	Medium Strength
	950	Seawater	>.5 Long 12.7 Wide Edge	Yes	1.0	<.5	Limited Data.	29	Medium Strength
13 Cr (.03C-12.8Cr-5Ni)	770	Water	?	No	4	?	High Frequency, Near Threshold Data. Magnitude of Small Crack Effect Enhanced at High R.	25	Medium Strength
Q1N (.17C-1.2Cr-2.4Ni)	625	Seawater	.5-6 Long 12.7 Wide Edge	Yes - Air No - NaCl	4	2.0	No Crack Size Effect at R = 0.5 or ΔK > 30 MPa√m. Limited Data for Air.	29, 30	Low Strength
BS4360:50D (.18C-1.2Mn)	370	Seawater	1-7 Long 23.5 Wide Edge	Yes - Air No - NaCl	3	1.0-3.0	Crack Size Effect Eliminated by Cathodic Polarization; da/dN Small to Large < 1.0.	31	Low Strength
EN5 (.3C-.7Mn)	300	Seawater	.5-6 Long 12.7 Wide Edge	? - Air No - NaCl	2	2.0	Limited Data.	29	Low Strength

Figure 25. Literature results demonstrating the effect of crack size on corrosion fatigue in steels exposed to aqueous environments (Gangloff and Wei, 1986)

Appendix II

Summary of "short" crack experimental work performed by Researchers

This section briefly summarizes the "short" crack experimental work including "short" crack test specimen geometry and the test techniques used by the researchers to characterize the behavior of "short" cracks of structural materials. Different types of specimen geometry such as

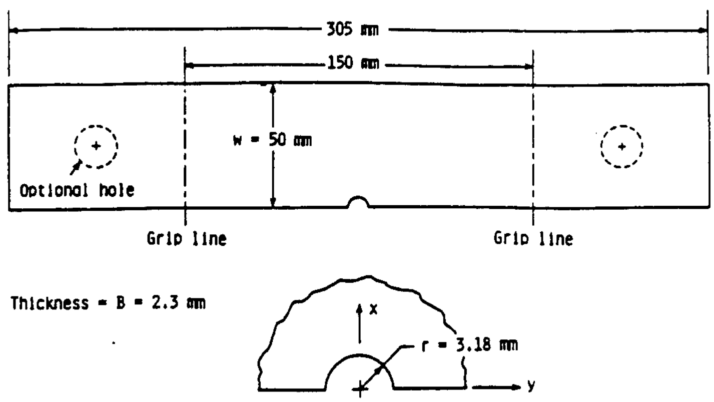
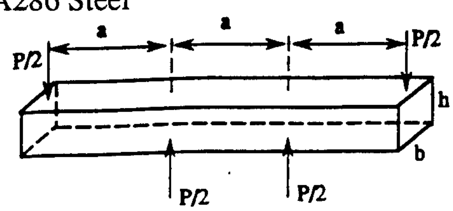
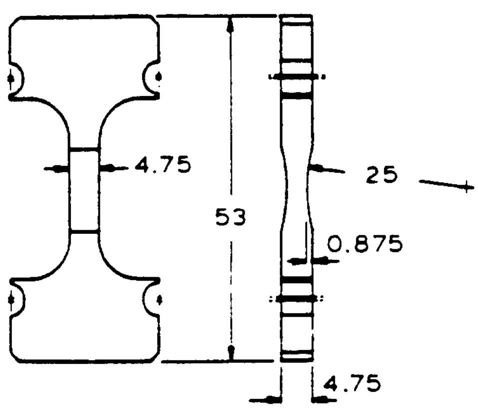
- Single-Edge-Notched-Tension (SENT) fatigue specimen,
- cylindrical "smooth" specimen,
- four point bend specimens, plate specimens,
- specimens notched by a hole, cut or center cracked tension (CCT),
- tensile square bar (TSB) type containing a shallow groove,
- shallow hour-glass shape specimens,
- specially designed "short" crack specimens with round or rectangular cross section,
- hollow cylindrical specimen, specimens with cross section of parallelogram, and
- specimens with reduced gage section have been used to study the behavior of "short" cracks.

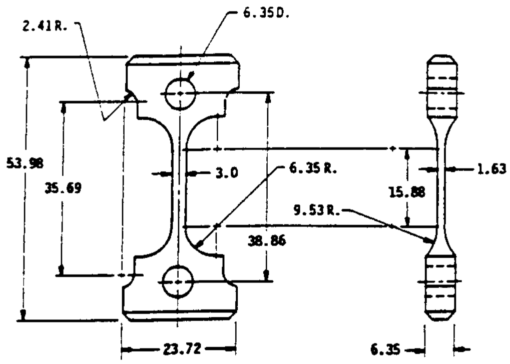
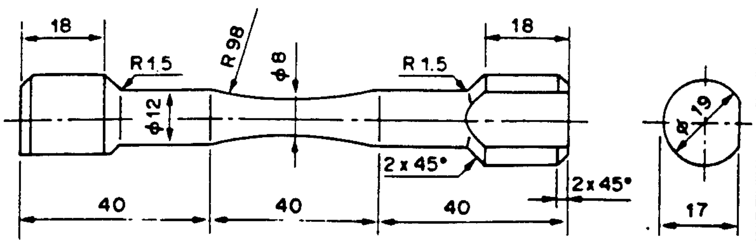
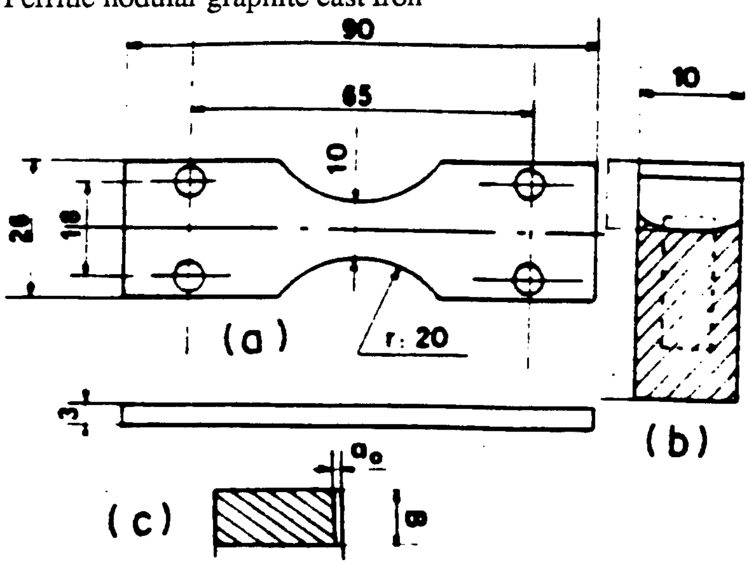
Various experimental techniques have been developed to observe the behavior of "short" cracks such as

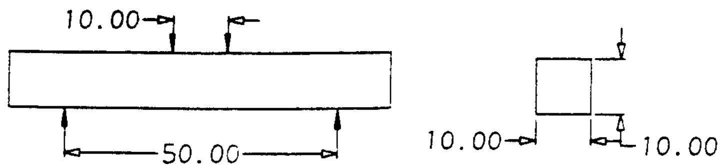
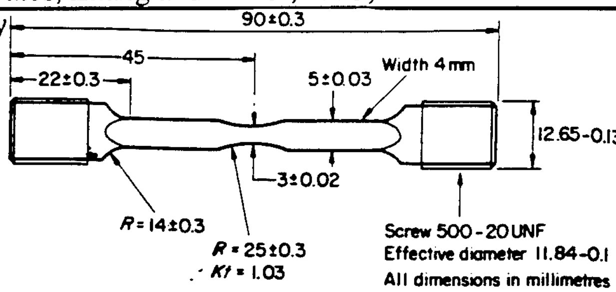
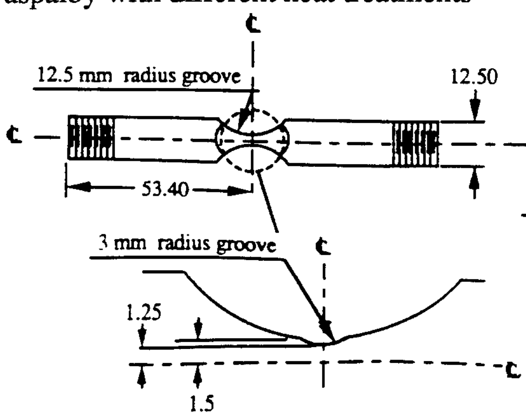
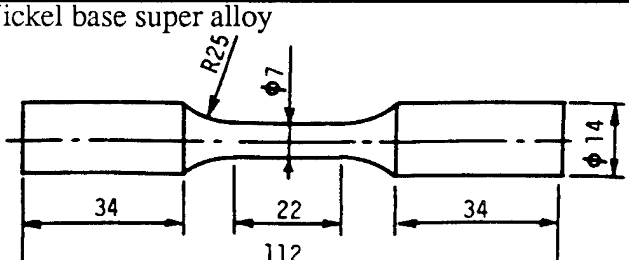
- photography,
- photomicroscopy,
- potential drop (AC & DC),
- compliance,
- metallography,
- replication,
- ultrasonic,

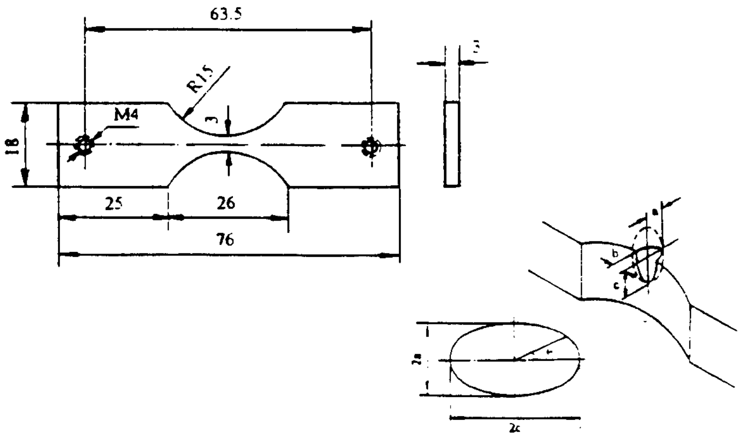
• In-situ techniques viz. Transmission Electron Microscope (TEM), Scanning Electron Microscope (SEM) and Scanning Laser Acoustic Microscope (SLAM).

Among these test techniques plastic replica and potential drop methods have been widely used. A few studies have used in-situ technique to observe the "short" crack behavior of materials. The in-situ techniques have significant advantage over the conventional methods, as the complete event of crack formation and growth can be monitored and recorded throughout the fatigue testing. Some researchers also have developed automatic crack monitoring system incorporating a camera and a microscope to capture the formation and the growth of "short" cracks at a predetermined fatigue cycles. A brief summary of "short" crack experimental works including the geometry of the "short" crack test specimens, the test conditions as well as the techniques used by several researchers is given in Table I.

Material/Specimen Geometry	Test Parameters & Test Technique
2024-T3 Aluminum Alloy  Single-edge notched tension (SENT) fatigue specimen (Newman and Edwards, 1988)	Loading-Constant Amplitude @ R=-2, -1, 0, 0.5 and @ FALSTAFF and GAUSSIAN. Maximum Gross Stress (MPa)- 75, 60 and 50 @ R=-2; 105, 80 and 70 @ R=-1; 145, 120 and 110 @ R=0; 225, 205 and 195 @ R=0.5, 275, 205 and 170 @ FALSTAFF; 170, 145 and 125 @ GAUSSIAN. Frequency-5-10 Hz Laboratory Air Plastic Replica
High strength A286 Steel  a = 21.2 mm b = 12.4 mm h = 9.4 mm Four-point bending specimen used for "short" fatigue crack "initiation" and propagation tests (Mei and Morris, 1993)	Loading-@0.5Sigma$_y$ (σ_{max}=571.8 MPa, σ_{min}=39.8 MPa); R=0.7; @0.8σ_y (σ_{max}=874 MPa; R=0.056) Laboratory air Optical microscope
7075-T6 Aluminum Alloy Specimen with reduced gage section with K_T=1.06  Design of "small" crack specimen-dimensions in mm (Lankford, 1982)	Loading-Constant Amplitude S_{max}=414 MPa (80% of Yield stress) @ R=0.05, Max. cyclic stress intensity = 414 MPa (80% of yield stress) Frequency=5Hz Laboratory air with RH=60% Test Technique Replica

Titanium-Aluminide alloy  Specimen design and dimensions (mm) used for the "initiation", growth and analysis of "small" fatigue cracks (Davidson, Campbell and Page, 1991)	Loaded in three-point bending; Loading- σ=1406 MPa $N_i^{0.069}$; R=0.1. Laboratory air Replica
Normalized medium carbon steel (0.4%C steel)  Shallow hour-glass shape specimen Specimen geometry-dimensions in mm (De los Rios, Tang and Miller, 1984)	Torsional deflection-controlled fatigue test; Stress level-350-400 MPa; Total strain range-0.5 to 0.65. Room Temperature Replica
Ferritic nodular graphite cast iron  Specimen dimensions in mm. (a) Uncracked specimens (b) Three point bend specimens. Solid lines for large specimens, dotted lines for "small" crack specimens (c) "short" crack specimens (Clement, Angeli and Pineau, 1984)	Four-point bending@R=0.1; Frequency=20-50 Hz. Laboratory air d.c. potential drop technique <u>"Short" crack formation procedure</u> Fatigue precracking of three-point bend specimens at a low ΔK approaching the near-threshold regime. After this the upper part and lateral faces of the specimens were machined as shown in fig. (b). With this procedure "short" crack in the order of a=0.075-0.5 mm was obtained with a "straight front". Specimens were heat treated to relieve the residual stresses because of the above procedure.

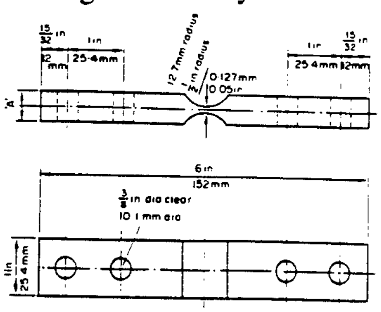
Cast and wrought nickel base super alloy-Waspaloy  Four point bend specimens and loading states-dimensions in mm (Yates, Zhang and Miller, 1993)	Four point bending tests @ frequency 30 Hz. S_{max}= 880 MPa @ R=0.1 waveform = sinusoidal. Acetate Replica
Waspaloy  "Short" fatigue crack specimen with shallow notch - K_t = 1.03 (Healy, Grabowski and Beevers, 1991)	Constant Amplitude @ R=0.5 and 0.1 at 19°C and 500°C. Constant load range @ 75 to 80% of proof stress. Frequency=100 Hz. Optical system and the use of Image analyzer.
Waspaloy with different heat treatments  Specimen geometry-all dimensions in mm (Stephens, Grabowski and Hoeppner, 1993)	Constant amplitude in load control A maximum stress of 92% of the 0.2% proof stress was applied. R=0.1 @ Frequency=20 Hz. At 25, 500, and 700°C. Insitu SEM in vacuum environment (10^{-4} torr)
Nickel base super alloy  Geometry of specimen used - dimensions in mm (Okazaki, Tabata and Nohmi, 1990)	Strain controlled low-cycle fatigue tests Temperature = 873°K The applied strain ratio was zero Frequency = 10 Hz. Acetate replica technique

<table>
<tr>
<td>8090 Aluminum-Lithium alloy


Plain and with corner notch -- Specimen size in mm and the corner notch profile (Pan, De los Rios, and Miller, 1993)</td>
<td>Tested at 245.8, 259.3, 279.4 and 241.4 MPa

R=0; Frequency=2Hz.

In-situ with an optical or acoustic system

Laboratory air</td>
</tr>
<tr>
<td>Aluminum-copper-magnesium alloy BS L65 and Aluminum-zinc-magnesium alloy DTD 5050

Test specimen for investigating fatigue crack initiation at a plane polished surface (Pearson, 1975)</td>
<td>Bending test @ fixed amplitude of 1500 cycles/min.

A microscope fitted with calibrated eyepiece.</td>
</tr>
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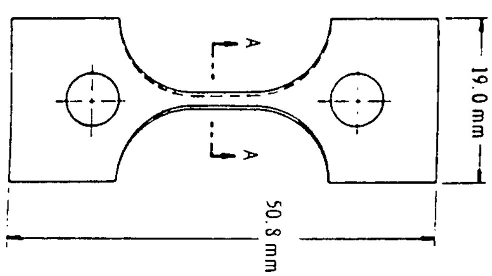
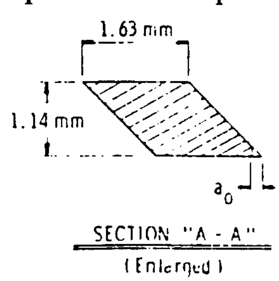
<td>Powder metallurgy nickel-base super alloy


Fatigue specimen design (Lankford, Cook and Sheldon, 1981)</td>
<td>Frequency 5 Hz

Laboratory air

Optical Microscope

</td>
</tr>
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<table>
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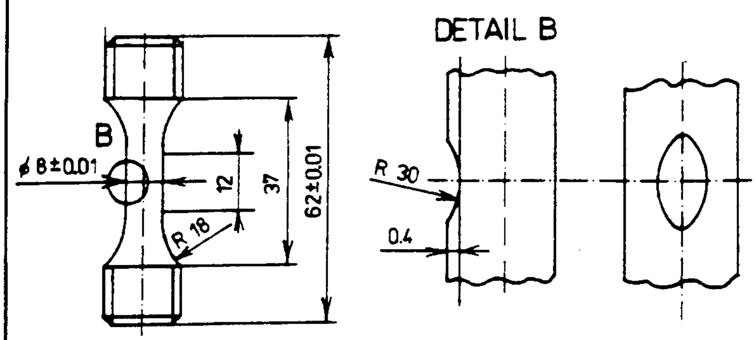
<td>Commercial steel sheet of grade 15Ch2NMFAA.


Specimen bar with shallow notch for investigating the initiation and propagation of short fatigue cracks (Hyspecky and Strnadel, 1992)</td>
<td>Tested @ two plastic deformation amplitude levels, $\varepsilon/2=0.003$ and $\varepsilon/2=0.005$ @ 20°C.

R = -1

Frequency=1 Hz.

SEM and optical microscope</td>
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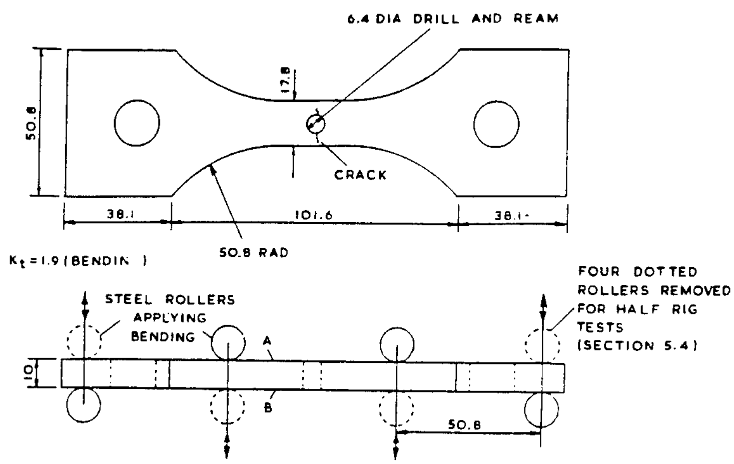
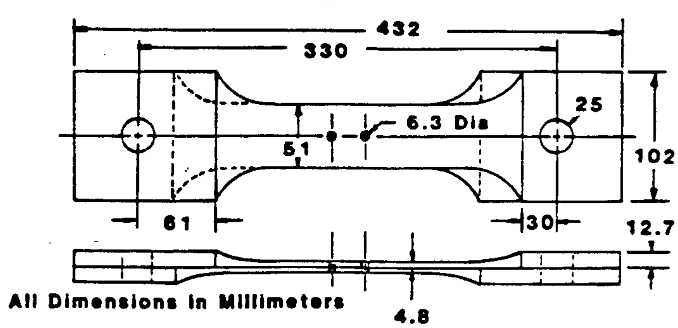
<td>4% Cu-Al alloy (BS 2L65)


Notched specimen (Cook and Edwards, 1982)</td>
<td>Four point bending

Variable amplitude loading used were Gaussian random and FALSTAFF

Replica</td>
</tr>
<tr>
<td>7475-T7651 aluminum alloy


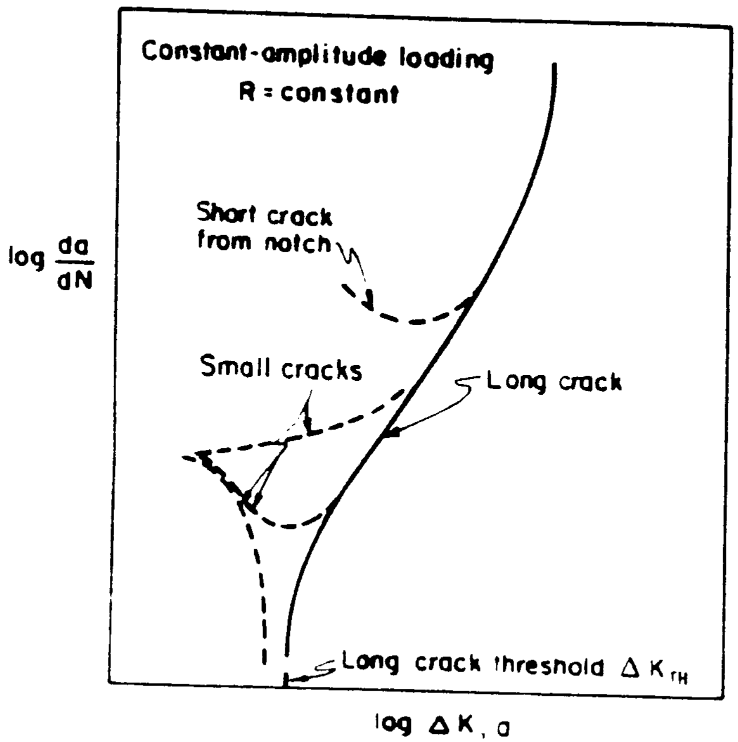
Test specimen geometry (Potter and Yee, 1982)</td>
<td>Fatigue tested to 16000 design usage flight hours.

Maximum gross section stress was 235 MPa.

Fractographic crack length measurement.</td>
</tr>
</table>

Appendix III

A global view of "short" crack challenge in materials



SCHEMATIC REPRESENTATION OF "SHORT" AND LONG CRACK GROWTH BEHAVIOR (Ritchie and Lankford, 1986)

TYPES

- Microstructurally "short"
  (cracks comparable in size with the scale of microstructural features such as grain size)
- Mechanically "short"
  (crack size comparable to the scale of local plasticity. e.g. crack tip plastic zone, strain field of a notch)
- Physically "short"
  (cracks typically less than 0.5 - 1 mm)
- Chemically "short"
  (size range depends on frequency and reaction kinetics)

LIMITATIONS

- LEFM not applicable -- No more "small scale" yielding
- No linear relationship between da/dN and Stress intensity range
- Continuum mechanics concepts are not valid.
- Breakdown in metallurgical similitude
- Non-availability of NDI techniques in "short" crack range.

CHARACTERISTIC FEATURES

- "faster" growth compared to long crack at equivalent or even less stress intensity range.
- growth rate often gets decreased or arrested at the grain boundary
- "microcrack" coalescence

CRACK DRIVING FORCE"

Mechanically "short" -- EPFM (Delta J)

Physically "short" -- closure-corrected stress intensity range

Microstructurally "short"-- probabilistic approach

MECHANISMS

<u>Microstructurally "short"</u>

- crack tip shielding (limited "wake")
- enhanced crack tip plastic strains resulting in "faster" growth
- non-uniform growth because of grain boundary obstruction (pinning of slip bands originating from crack tip by grain boundary)
- crack deflection resulting from grain (non-favorable) orientation
- crack arrest results in non-propagating "short" cracks

<u>Mechanically "short"</u>

- Excessive plasticity

<u>Physically "short"</u>

- Crack tip shielding
- absence of crack closure

<u>Chemically "short"</u>

- Local crack tip environment

FACTORS INFLUENCE "SHORT" CRACK BEHAVIOR

- <u>Material Parameters</u>
  - grain size
  - grain orientation
  - texture
  - slip band character
  - local microscopic fracture toughness
  - inclusion size and content
  - second phase particles
  - thermomechanical treatment

- <u>Mechanical parameters</u>
  - applied stress
  - R value
  - load spectrum
  - specimen geometry
  - notch size
  - fabrication techniques (e.g. riveting)
  - processing techniques

- <u>Environment (chemical)</u>
  - frequency
  - reaction kinetics
  - hydrogen

- Temperature

"SHORT" CRACK FORMATION SITES IN MATERIALS TESTED

Aluminum alloys (Pearson, 1975)
BS L65 (Al-Cu-Mg) and DTD 5050 (Al-Zn-Mg)
- surface inclusions and second phase particles (pearson 1975)

In 2219-T851(Morris, Buck and Marcus, 1976)
- surface intermetallic inclusions (Morris et al, 1976)

In 8090 Al-Li alloy (Pan, Rios and Miller, 1993)
- inclusions
- corner notches

In SiC reinforced aluminum alloy (6061) (Kumai, King and Knott, 1990)
- at SiC particles

In 2024 aluminum alloy (Sigler, Montpetit and Haworth, 1983)
- at voids

In 2090-T8E41 Al-Li alloy (Rao, Yu and Ritchie, 1988)
- inclusions and intermetallic particles

In Al-Mg-Si alloy (Plumtree and O'Connor, 1991)
- second phase particles

In nodular cast iron (Clement, Angeli and Pineau, 1984)
- at graphite nodules
- microshrinkage pores

In medium carbon steel (Rios, Tang and Miller, 1984)
- ferrite-pearlite boundaries

In low carbon steel (Tokaji, Ogawa and Harada, 1986)
- in fine grained material within ferrite grains
- in coarse grained material, at grain boundaries

"SHORT" CRACK FORMATION SITES IN MATERIALS TESTED

In Ti-6Al-2Sn-4Zr-6Mo (Mahajan and Margolin, 1980)
- at higher strains within the slip bands
- at low strains at alpha-beta interface

In 2024 and 2124 aluminum alloy in the T4 condition (Kung and Fine, 1979)
- at higher stresses at "coarse slip lines"
- at low stresses from constituent particles

In polycrystalline copper (Polak and Liskutin, 1990)
- at slip bands within the grain as well as in preferably oriented grain boundaries

In Al-Mg-Si alloy extruded and squeeze-cast (Plumtree and O'Connor, 1991)
- slip band stage I

In Waspaloy (Ni base super alloy) (Yates, Zhang and Miller, 1993)
- from slip bands
- along a twin boundary
- in grains

In titanium-aluminide alloy (Davidson, Campbell and Page, 1991)
- from Ti_3Al hcp phase

In high and low alloy steel containing V and Nb (Kim and Fine, 1982)
- slip band cracking

In extruded aluminum alloy X7091 containing Zn-Mg-Cu (Hirose and Fine, 1983)
- at low stresses at grain boundaries
- at high stresses at slip bands

"SHORT" CRACK FORMATION SITES IN MATERIALS TESTED

TEMPERATURE EFFECTS

<u>In nickel base super alloy</u> (Mei, Krenn and Morris, 1993)
- at room temperature at inclusions and particles and less frequently at grain boundaries
- at 873°K at micropores, slip planes and carbide precipitates at grain boundaries

<u>In Waspaloy</u> (Stephens, Grobowski and Hoeppner, 1993)
- at 25 and 500°C at twin boundaries
- at 700°C at slip bands

<u>In Waspaloy</u> (Healy, Grabowski and Beevers, 1991)
- at 19 and 500°C at coarse carbide particles

<u>In 304 stainless steel</u> (Suh, Lee and Kang, 1990)
- at 538°C at grain boundaries

ENVIRONMENT EFFECTS

<u>In age hardened Al-Zn-Mg alloy</u>
(Nageswararao and Gerold, 1976)

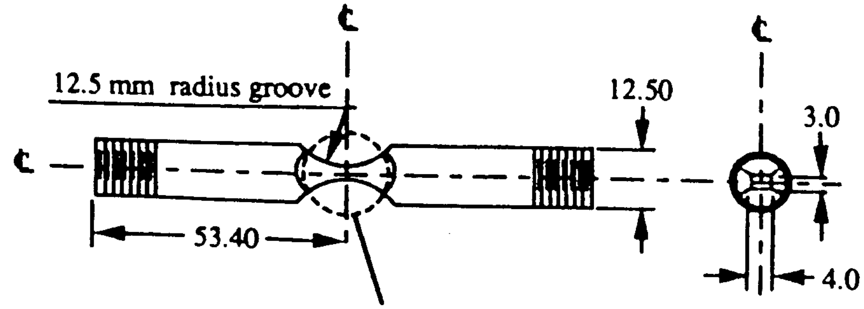
- in dry nitrogen gas environment along favorably oriented crystallographic directions

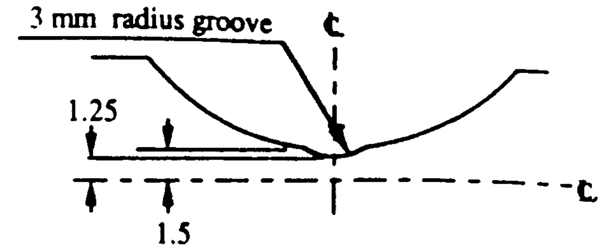
<u>In high purity polycrystalline copper</u>
(Masuda and Duquette, 1975)

- in laboratory air as well as in 0.5% NaCl solution, at slip bands

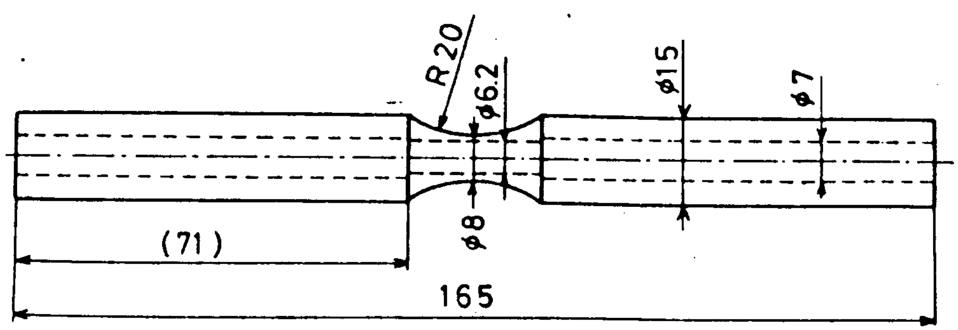
"SHORT" CRACK TEST SPECIMEN GEOMETRY

- Single-Edge-Notched-Tension (SENT) fatigue specimens
- cylindrical "smooth" specimen
- four point bend specimen
- plate specimens
- specimens notched by a hole, cut or center cracked tension (CCT) specimen
- tensile square bar (TSB) type containing a shallow groove
- shallow hour-glass shape specimen
- specimens with round or rectangular cross section
- hollow cylindrical specimen
- specimens with cross section of parallelogram and reduced gage section

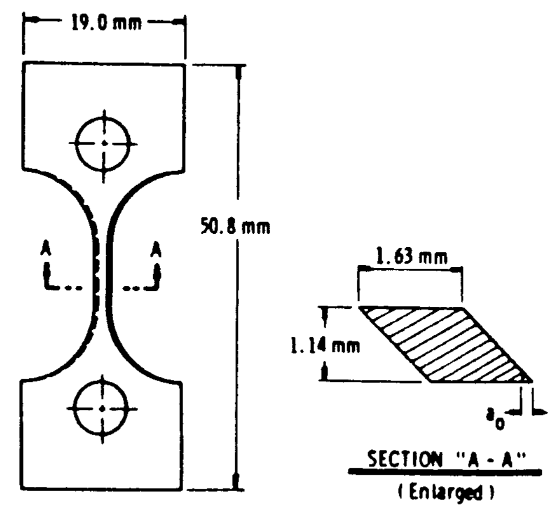




Stephens, Grabowski and Hoeppner, 1993

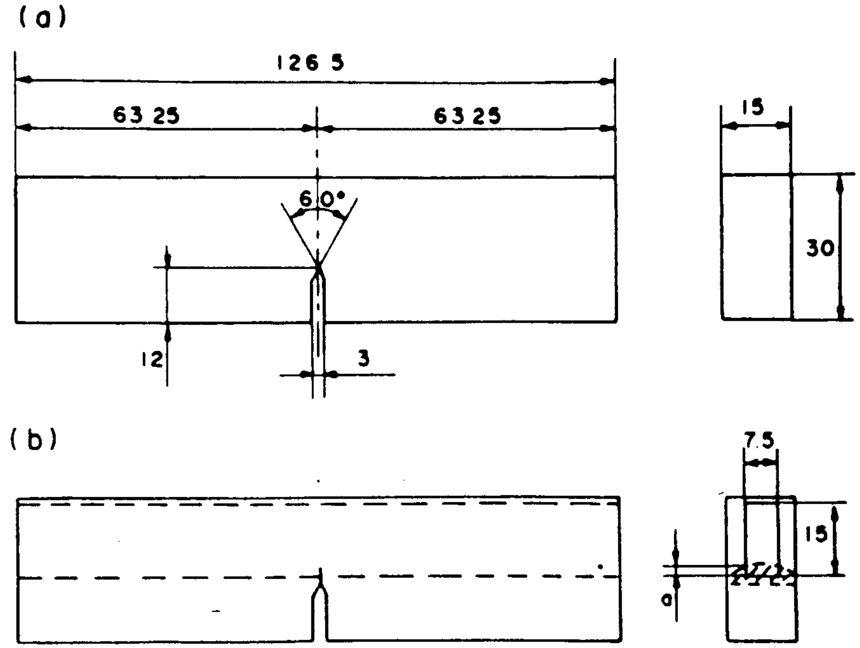


Kunio et al, 1981

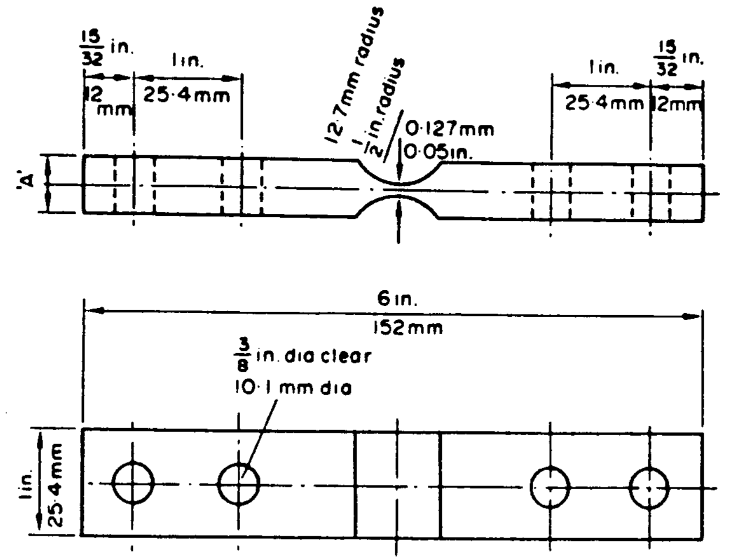


Sheldon, Cook, Jones, and Lankford, 1981

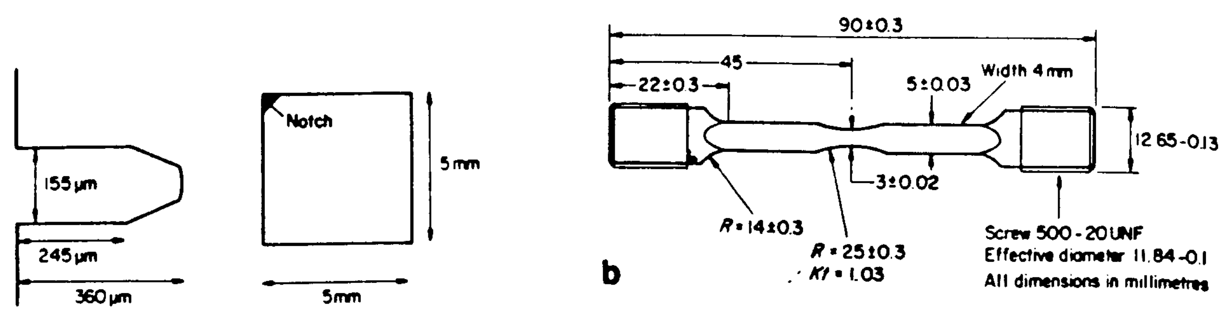
"SHORT" CRACK TEST SPECIMEN GEOMETRY



Breat, Mudry and Pineau, 1983

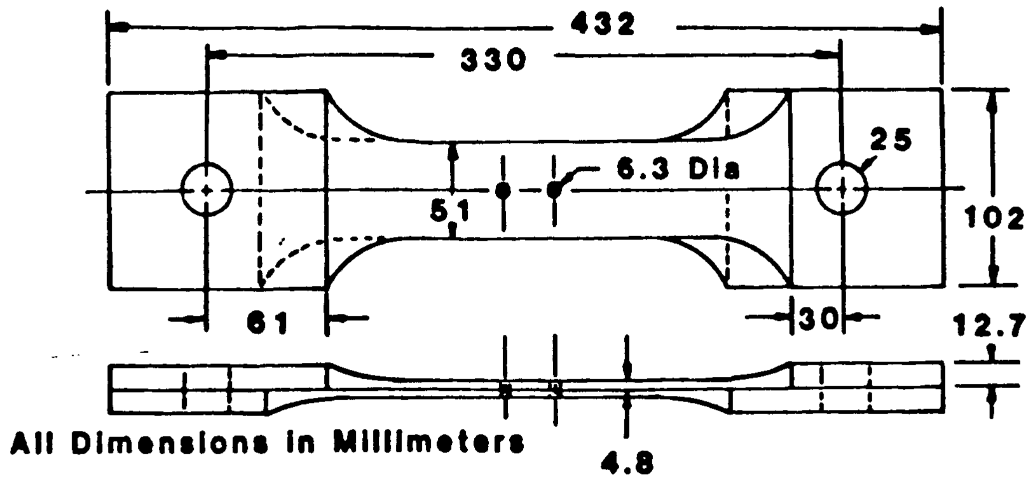


Pearson, 1975

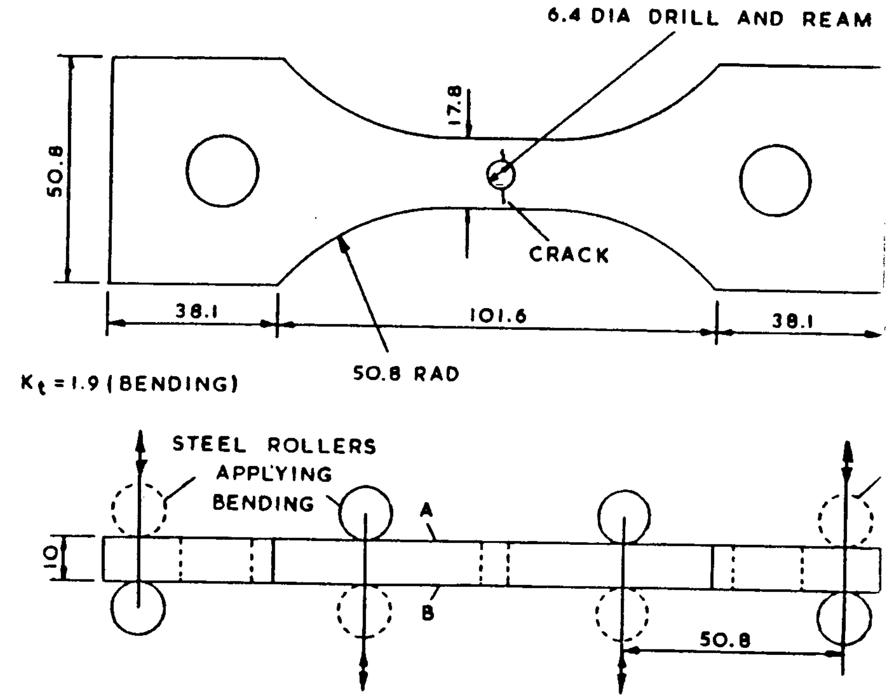


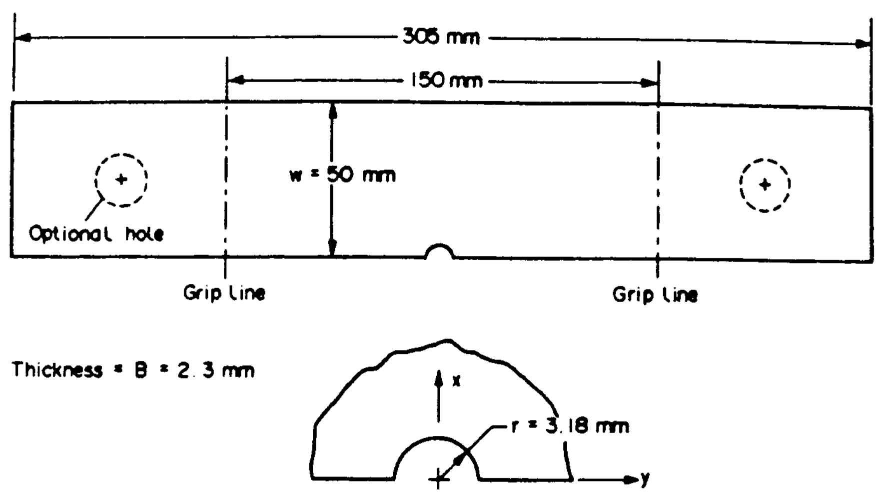
Healy, Grabowski and Beevers, 1991

"SHORT" CRACK TEST SPECIMEN GEOMETRY



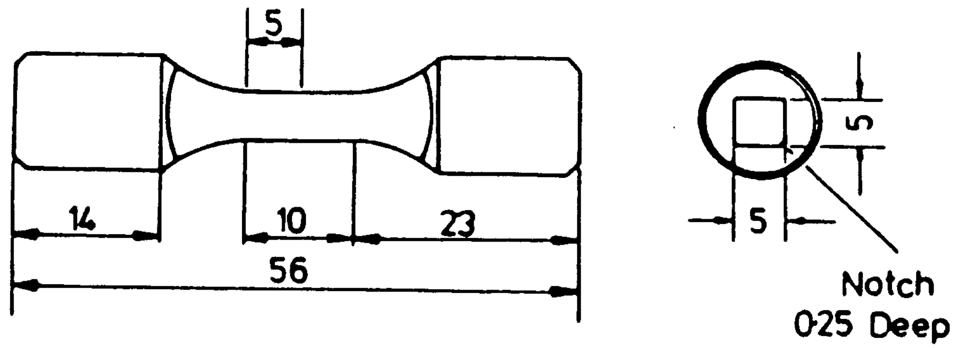
Potter and Yee, 1982



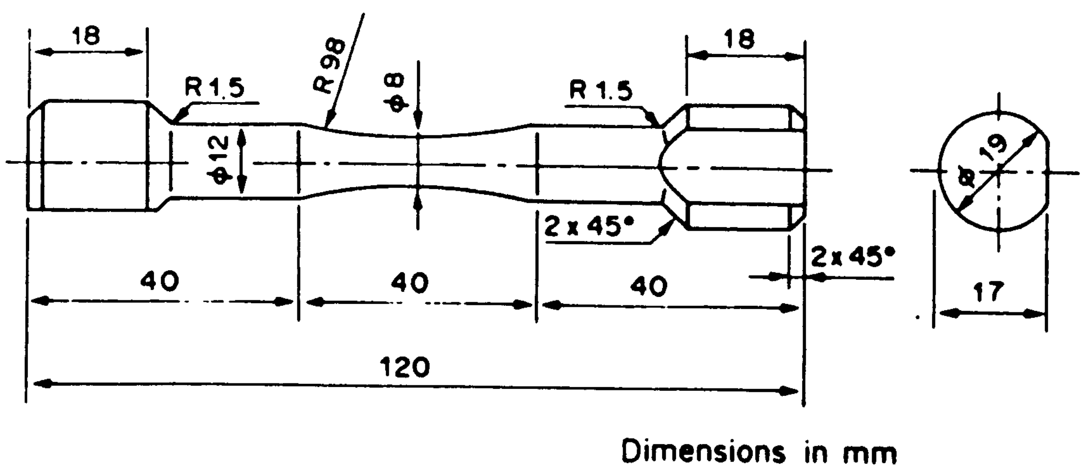


Newman and Edwards, 1988

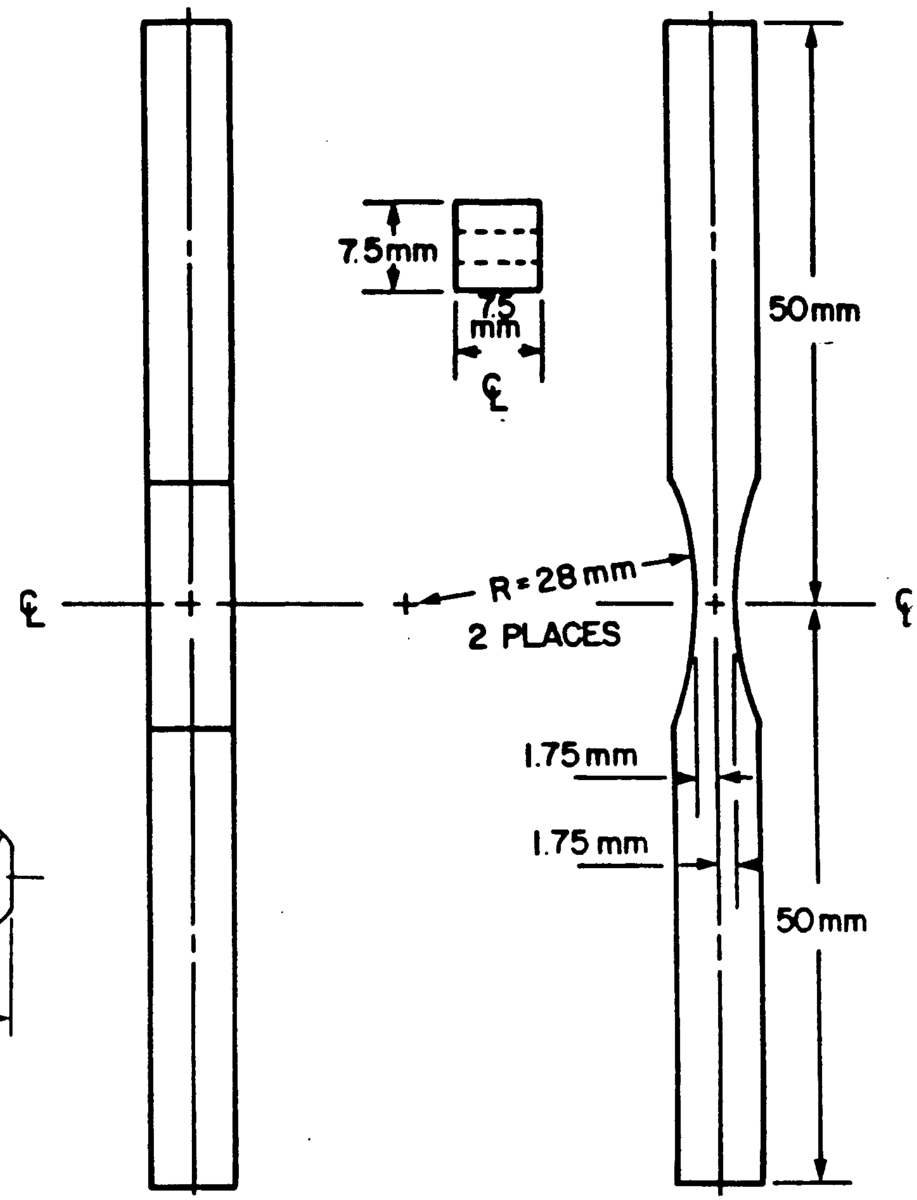
"SHORT" CRACK TEST SPECIMEN GEOMETRY



Howland, 1986

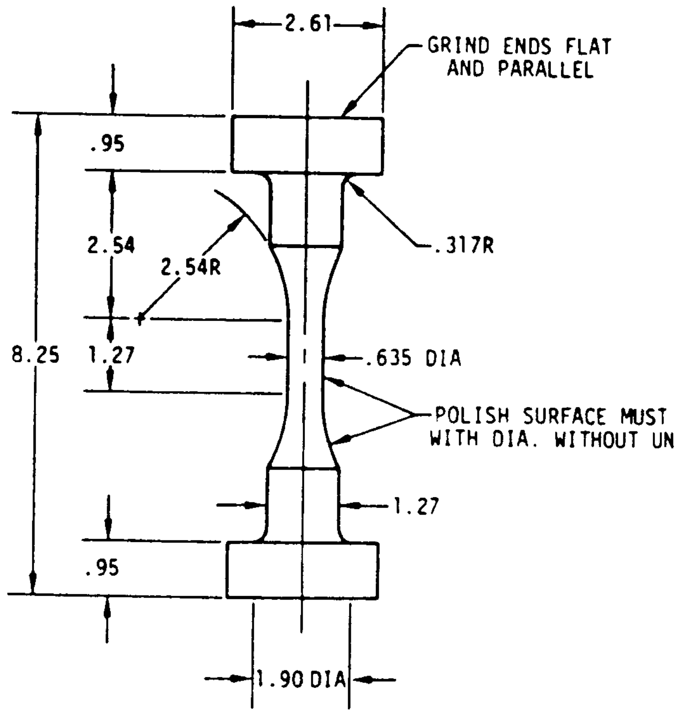


De los Rios, Tang and Miller, 1984

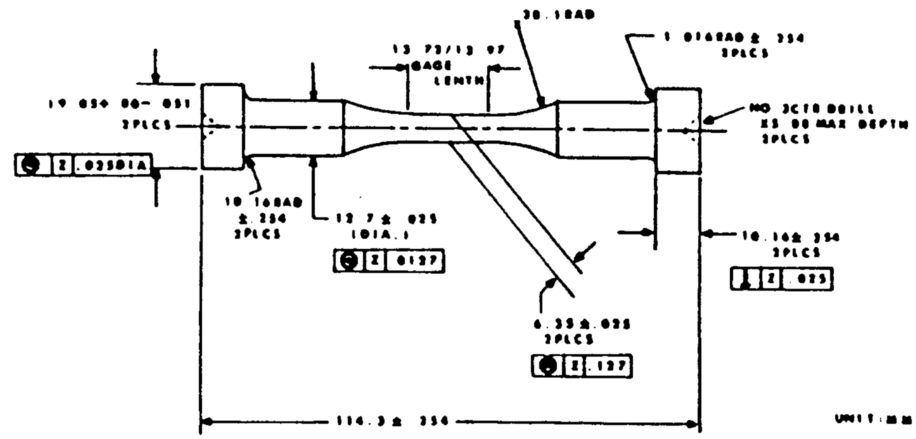


Larsen et al, 1986; Miller, 1984

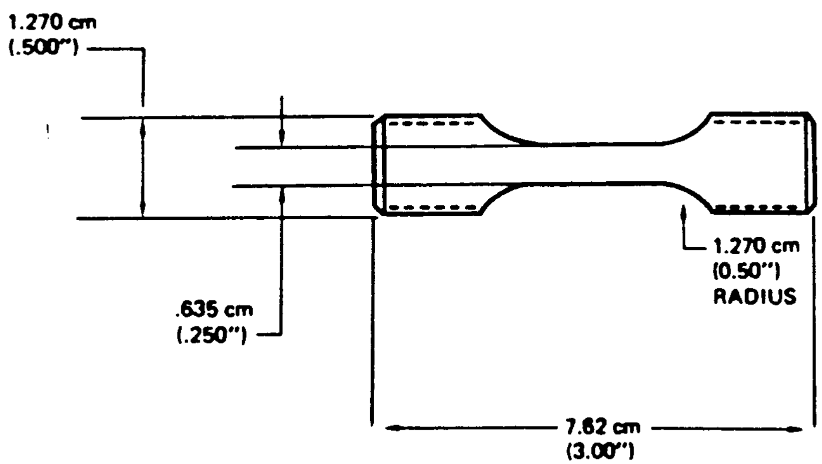
"SHORT" CRACK TEST SPECIMEN GEOMETRY



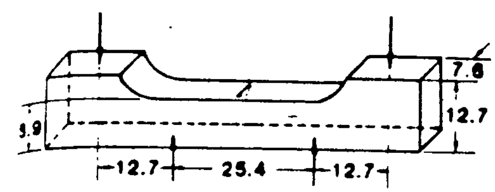
Morris, 1977



Mahajan and Margolin, 1982

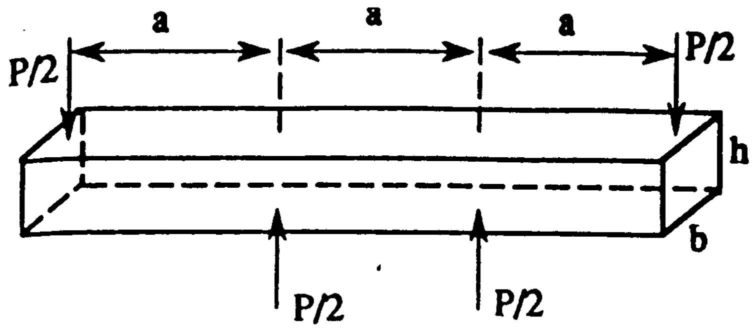


Hyzak and Bernstein, 1982



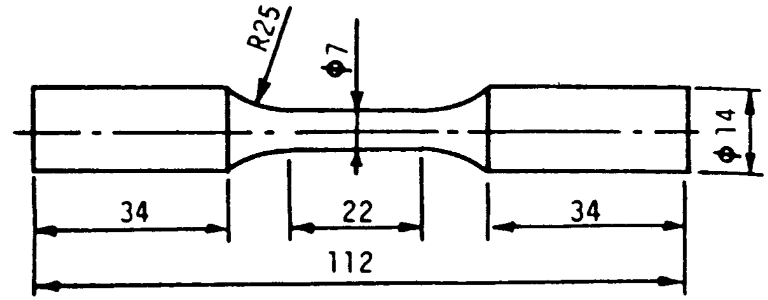
Venkateswara Rao, Yu and Ritchie, 1988

"SHORT" CRACK TEST SPECIMEN GEOMETRY

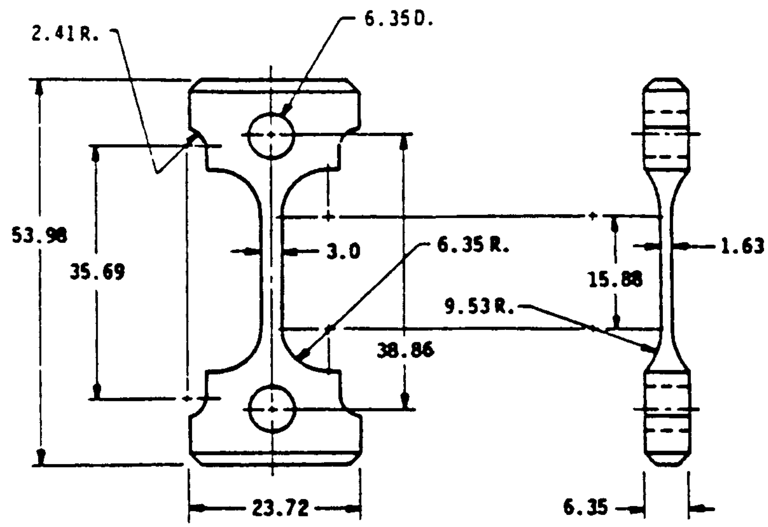


a = 21.2 mm b = 12.4 mm h = 9.4 mm

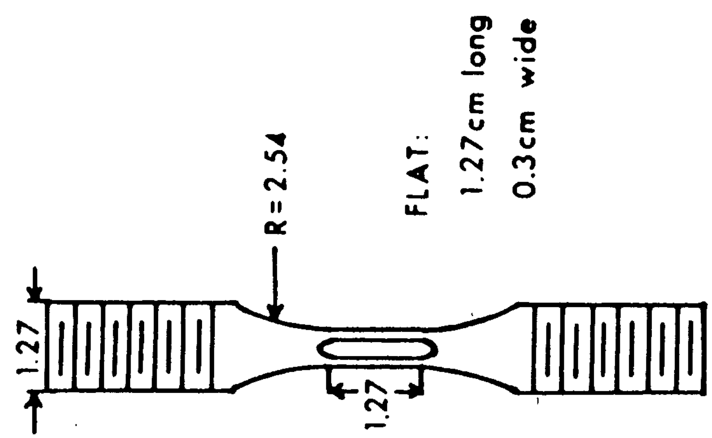
Mei and Morris, 1993



Okazaki, Tabata and Nohmi, 1990



Davidson, Campbell and Page, 1991



Sigler, Montpetit and Haworth, 1983

"SHORT" CRACK EXPERIMENTAL TECHNIQUES

- photography
- photomicroscopy
- AC and DC potential drop
- metallography
- fractography
- compliance
- replication
- ultrasonic Rayleigh wave technique
- in-situ techniques using SEM and TEM
- electro-chemical detection
- automated photomicroscopic system

"SHORT" CRACK MODELLING METHODS

Physical models

- dislocation based model (modeling the crack as an array of dislocations and the resolved shear stress as the crack driving force) (Estabrook, 1984)
- model relating da/dN, crack length and the distance between the crack tip and the nearest grain boundary (Hobson, 1982)
- analytical crack closure model (Newman, 1983) (approximately characterizes the da/dN by LEFM parameters into "short" crack regime).

Empirical Models

- model incorporating microstructural influences (Grabowski and King, 1992)
- microcrack propagation model (Morris, James and Buck, 1981)
- Plasticity induced crack-closure model (Newman, 1994)
- Microstructural barrier model (Miller et al, 1985)

Appendix IV
Title list of literature related to "short" crack studies

EFFECT OF MICROSTRUCTURE

W.J. Baxter (1986), "The growth of persistent slip bands during fatigue", The Behavior of Short Fatigue Cracks, K.J. Miller and E.R. de los Rios, Eds, Mechanical Engineering Publications Limited, pp. 193-202.

R.K. Bolinbroke and J.E. King (1986), "The growth of short fatigue cracks in titanium alloys IMI550 and IMI318", Small Fatigue Cracks, R.O. Ritchie and J. Lankford, Eds., A Publication of The Metallurgical Society, Inc., pp. 129-144.

C.W. Brown, J.E. King, and M.A. Hicks (1984), "Effects of microstructure on long and short crack growth in nickel base super alloys", Metal Science, 18, pp. 374-380.

C.W. Brown and D. Taylor (1984), "The effects of texture and grain size on the short fatigue crack growth rates in Ti-6Al-4V", Fatigue Crack Threshold Concepts, D.L. Davidson and S. Suresh, Eds, AIME, pp. 433-446.

C.W. Brown and J.E. King (1986), "The relevance of microstructural influences in the short crack regime to overall fatigue resistance", Small Fatigue Cracks, R.O. Ritchie and J. Lankford, Eds., A Publication of The Metallurgical Society, Inc., pp. 73-96.

K.S. Chan and J. Lankford (1988), "The role of microstructural dissimilitude in fatigue and fracture of small cracks", Acta Metallurgica, pp. 193-206.

P. Clement, J.P. Angeli and A. Pineau (1984), "Short crack behavior in nodular cast iron", Fatigue Fracture of Engineering Materials and Structures, 7, pp. 251-265.

M.A. Daeubler and A.W. Thompson (1986), "Microstructural effects on short fatigue cracks in eutectoid steels", Small Fatigue Cracks, R.O. Ritchie and J. Lankford, Eds., A Publication of The Metallurgical Society, Inc., pp. 157-166.

D.L. Davidson (1988), "Small and large fatigue cracks in aluminum alloys", Acta Metallurgica, 36, pp. 2275-2282.

D.L. Davidson, J.B. Campbell, and R.A. Page (1991), "The initiation and growth of fatigue cracks in a titanium aluminide alloy", Metallurgical Transactions, 22A, pp. 377-391.

E.R. de los Rios, Z. Tang and K.J. Miller (1984), "Short crack fatigue behavior in a medium carbon steel", Fatigue Fracture of Engineering Materials and Structures, 7, pp. 97-108, 1984.

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